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ADVISORY GROUP FOR AEROSPACE RESEARCH & DEVELOPMENT

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Maintenance in Service of High Temperature Parts

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NORTH ATLANTIC TREATY ORG   IZATION

ADVISORY GROUP FOR AEROSPACE RESEARCH AND DEVELOPMENT

(ORGANISATION DU TRAITE DE L'ATLANTIQUE NORD)

AGARD Conference Proceedings No.317

MAINTENANCE IN SERVICE OF HIGH-TEMPERATURE PARTS

Papers presented at the 53rd Meeting of the AGARD Structures and Materials Panel held in Noordwijkerhout, the Netherlands 27 September–2 October 1981.

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PREFACE

Over the years the Structures and Materials Panel has been concerned with the problems of gas turbine engine materials. The Panel sponsored a co-operative testing programme to analyse and predict low cycle high temperature fatigue life and this was the topic for a Specialists' Meeting, namely:

> "Characterisation of Low Cycle High Temperature Fatigue by the Strain Range Partitioning Method" (Aalborg 1978, AGARD CP-243).

Another approach was the use of ceramics, discussed at a Specialists' Meeting entitled:

> "Ceramics for Gas Turbine Applications" (Porz-Wahn (Cologne) 1979, AGARD CP-276).

While it was felt that these Meetings had achieved their purpose for their specialist audiences, there remained a gap to be bridged between the experience and information of the maintenance engineer and what the materials specialist in the research laboratory could offer. This Specialists' Meeting on the "Maintenance in Service of High Temperature Parts" was conceived as a two way exchange between maintenance experience and materials science.

It was felt that AGARD should be involved because all NATO countries faced common problems in the increasing costs of engine maintenance and the scarcity of strategic materials. Many of the problem areas were likely to have a common base in relation to service experience and materials behaviour characteristics, so that an exchange of views should benefit both the research and maintenance communities. The collected proceedings and the discussion sessions endeavour to provide such an advantage to those involved in materials for gas turbine engines.

On behalf of the Structures and Materials Panel I would like to express my thanks to all authors, recorders, session chairmen and participants. In particular I appreciated the help of the members of the Sub-Committee on the Maintenance of High Temperature Parts when I became Chairman and that of Mr George C. Deutsch who, though retired, returned to conduct the final technical and discussion session of the Meeting.

D.A.FANNER
Chairman – Sub-Committee on
Maintenance of High Temperature Parts

CONTENTS

INTRODUCTORY REMARKS

by the

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The present Specialists' Meeting on the Maintenance in Service of High-Temperature Parts is not a routine undertaking in the history of the AGARD Structures and Materials Panel. Traditionally the SMP has concentrated on problems related to the airframe rather than the engine.

Even at its first meeting in Ottawa, June 1955, when it discussed high-temperature problems, this was in relation to a proposed AGARDograph on High-Temperature Effects on Aircraft Structures. At the 5th Panel Meeting in Oslo and Copenhagen, April/May 1957, Materials for Use at Elevated Temperatures were the main topic, but again the lectures that were held had basically the airframe in mind; this in spite of the fact that considerable attention was paid to titanium and its alloys.

Yet at certain intervals we have occupied ourselves with topics that were of interest to the engine community. Since so many representatives of that community are present now, I would like to tell you some more of our activities in that field. For your information I have also prepared a list of Engine Oriented Publications under the Auspices of the AGARD Structures and Materials Panel.

The Summary Record of the 5th Panel Meeting contained an Appendix called the Scope of the Work of the Panel. Here we find formally listed as topic 3h: The properties of aircraft and *aero-engine materials*, both metallic and non-metallic, including synthetics, adhesives and plastics.

This resulted in the following papers being presented at the Technical Sessions on Fatigue at the occasion of the 6th Panel Meeting in Paris, November 1957: Fatigue of Structural Materials at High Temperatures, by Prof. B.J.Lazan, University of Minnesota, and Corrosion Sèche et Protection des Alliages Réfractaires Ni-Cr 80/20, by M.Mathieu, ONERA.

At that meeting a Permanent Materials Committee was formed.

This committee stimulated activities in the high temperature materials field, not so much with an eye on present day problems, but much more in relation to future developments.

Topics treated included:

- Cermets, 7th Panel Meeting, Paris, March/April 1958,
- Unconventional Metals (Be, Cr, Mo, Nb, Ta, Ti, W and V), first suggested at the 8th Panel Meeting, Paris, October 1958.

At the 9th Panel Meeting, Paris, April 1959, the SMP met in its re-organized form, incorporating a Structures Group, a Materials Group and a Flutter Committee.

The Materials Group initiated an activity which was to continue for a considerable number of years and which was devoted to the four prominent refractory metals: Mo, Nb, Ta and W and their coatings. Several aspects were considered:

1. Alloy development
2. Coatings
3. Production development
4. Welding
5. Basic Research
   (phase diagrams, diffusion studies, impurity effects, new fabrication techniques, oxidation studies)
6. Design data.

Various publications resulted, but a major achievement in my mind was a handbook on the impurity analysis of the four refractory metals, which was the result of a collaborative study that lasted for many years.

Another topic that was pursued was that of graphite as a high temperature structural material.

High temperature mechanical and creep testing were the subjects of two other collaborative efforts.

At the 19th Panel Meeting, Paris, October 1964, it was for the first time that a topic was taken up, that was of immediate interest to the designers and the users of jet engines. This was the development and testing of protective coatings for jet engine materials. Unfortunately this did not result in an AGARD-sponsored collaborative programme. Presumably this was because several potential participants did not feel their commercial interests sufficiently safe-guarded.

It is important of course that there is another AGARD Panel especially devoted to engines and their problems. This is the Propulsion and Energetics Panel. In 1965 the SMP was invited to contribute to a PEP meeting on engine materials. Since that time liaison has been maintained with the PEP and the two Panels have assisted one another in soliciting speakers for their respective conferences.

In the years that followed attention was again given to the more basic aspects such as the thermo-physical properties of materials.

The 28th Panel Meeting, Dayton, Ohio, April/May 1969 marks a sudden growth in the Panel's interest in engine problems. High Temperature Corrosion was selected as a topic worthy of Panel attention.

This resulted in a Directory of Organizations, Investigators and Programmes in High-Temperature Corrosion Research completed in 1971, in a highly successful Specialists' Meeting on High Temperature Corrosion which took place in Copenhagen in the spring of 1972 and in a Handbook on Basic Data in High-Temperature Corrosion completed in 1974.

Another engine-related topic considered by the Panel was High-Temperature Low-Cycle Fatigue. First a state-of-the-art report was prepared on testing techniques and methods of analysis. Then a Specialists' Meeting was held in the spring of 1974.

In parallel to the HTLCF project an activity was undertaken concerning In-Situ Composites. This programme was run similarly to that on high-temperature corrosion. First a directory was prepared listing workers and organizations active in research and development in this field. Then a Specialists' Meeting on In-Situ Composites was held also at the occasion of the SMP 1974 Spring meeting in Washington D.C.

The topic of high-temperature design problems was held alive by an ad hoc group which, after listening to a number of pilot papers, proposed another activity in the field of low-cycle fatigue. This was to comprise a collaborative testing programme with as one of its aims to evaluate the strain-range partitioning method for the analysis and prediction of low cycle fatigue behaviour of engine materials. Sixteen laboratories participated, equally distributed to both sides of the Atlantic. The results of the testing programme were discussed at a Specialists' Meeting that took place in Aalborg in the Spring of 1978.

In parallel to the Low-Cycle-Fatigue Programme an activity was started in the field of ceramics. Speakers were invited to present pilot papers and in the Autumn of 1979 in Cologne, a Specialists' Meeting was held on Ceramics for Turbine Engine Applications.

At the 48th Panel Meeting in Williamsburg Va. April 1979, consideration was given to a proposal to stimulate the dialogue between engine maintenance engineers and materials specialists.

At the 49th Panel Meeting in Cologne, October 1979, a proposal was made to prepare a Handbook on Aeroelasticity for Turbomachines.

Discussions continued on the aspects of engine maintenance and these have led to the organisation of the Specialists' Meeting on Maintenance in Service of High-Temperature Parts to which you have been invited to participate. As you will see various categories of papers will be presented.

In the first category fall the papers that deal with the users' experience, both military and civil, with inspection, maintenance and repair of jet engines.

Another category discusses defects and criteria for engine life and cause for retirement.

A third category considers novel repair techniques and rejuvenation of parts that have been in service.

Finally a fourth category is concerned with protective coatings, to be applied in the manufacturing stage, but also at regular intervals during service life.

We have aimed at bringing together people with strongly different backgrounds, but all in some way involved in the problem of providing present day aircraft with reliable engines, that will continue to be reliable for an appreciable length of time.

By doing so we have aimed not only at the nations that produce engines, but also at those that have to operate and maintain engines produced elsewhere.

The engine environment is characterised by heat and pressure. Maybe the discussions during the Specialists' Meeting will create another form of heat and pressure. But then, it has been said that sufficient heat and pressure can turn carbon into diamond. I am not expecting that after this Meeting the floor will be found strewn with diamonds, but I am convinced that the results achieved will be valuable for some time to come; and I wish you all a pleasant and fruitful meeting.

H.P. VAN LEEUWEN

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MILITARY MAINTENANCE POLICIES AND PROCEDURES FOR HIGH-TEMPERATURE PARTS - WILL THEY BE ADEQUATE?

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SUMMARY

Compared with Civil use, military aero-engines tend to be exposed to the effects of working closer to their design limits, of more rapid and frequent change of operating condition, and of more exacting environments. For high temperature parts this mode of use generates three areas of concern; estimating the life of critical parts, deterioration in material properties, and difficulties in inspection and repair.

Our lifing methods are already complicated by an economic desire to move away from average fleet lines to consider the life consumption of individual engines. Newer materials are aggravating this by the number of parameters, all varying in service, which now affect the life. There is nothing new in the maintenance engineer's ideal of materials which do not deteriorate and are simple to inspect and repair. However the conflict between this ideal and the demands for lighter weight and higher performance has probably never been greater. Will we be able to develop maintenance policies and procedures to maintain a sensible balance?

1. INTRODUCTION

From the beginning the physical properties of materials and particularly those of high-temperature materials have been one of the main controlling factors in the development of gas turbine aero-engines for military use. In fact, from a viewpoint back in the year 1940, miracles have been achieved. Much has been written on the subject and this paper is not intended to add to that aspect. It is biased more towards the problems and challenges that the use of modern high-temperature materials bring to the Royal Air Force operator; living, as we all do these days, within strict socio-economic restraints.

Material has to be considered in relation to its overall effect on the aero-engine as part of a weapon system. Moreover, the weapon system itself has to behave quite differently in peace and wartime conditions. The maintenance man is faced with problems of life estimation, with all its facets, inspection, repair and containment of costs; both direct and indirect. From all of these factors maintenance policies emerge. Each of these areas is touched on separately in this paper. Of course this is merely for editorial convenience because, in real life, they are inextricably connected in very complex relationships and what appears at the end is often a compromise between several aspects.

2. OPERATIONAL WEAPON SYSTEM EFFECTS

2.1 The Military Aero-Engine. The military gas turbine is probably unique amongst aircraft equipment. Few other items operate so closely to the limit of physical properties of materials, has to undergo such intensive development of reliability and life after it has entered service, has such a high proportion of critical and life-limited parts and is so frequently returned to the repair/overhaul organisation to be dismantled.

Then with each new aircraft type the demand for performance forces us towards engines that are smaller, lighter and use less fuel for a given thrust. If we compare the Rolls-Royce RB 199 for Tornado with an early Rolls Royce Avon of approximately equal dry thrust, advancing technology has achieved a significant reduction in the size of the basic turbo machinery. Even with its reheat/afterburner the RB 199 is no bigger than the basic Avon. To a large degree this change is due to the production of materials which will withstand higher turbine temperatures and to the design of improved cooling arrangements. Generally both are involved.

2.2 Peace-time and War-time Conditions. It is only during relatively short periods of flight that engines really consume their in-built life. Although aircraft, and hence engine performance is primarily specified for their war role, in practice peace-time training sorties can consume engine life more quickly per flying hour than war-time operations. Fig 1 shows where training sorties are condensed to eliminate transit flying and intensify training benefit.

Perhaps in peacetime we should make do with lower thrusts and hence improve the turbine environment. To some extent this can be done. For example with some aircraft we use a flexible take-off thrust. The thrust used is adjusted to take account of aircraft weight, weather and runway conditions. However many pilots feel, with some justification, that the only good piece of runway is the portion that is left in front of you. For them, use of maximum thrust equates to improved aircraft safety at a very critical point in the sortie. Of course a more mechanical solution can be adopted and we do consider for new aircraft the fitting of "combat switches". These would limit thrust normally but provide a facility in the cockpit to obtain maximum thrust for realistic training or in an emergency.

2.3 Overall Maintenance Philosophy. Like many other aircraft operators, the RAF has tried many philosophies of maintenance in the past. We are now firmly settled on a philosophy of On Condition Maintenance. This is a controversial phrase and many people define it in different ways. More importantly, they also understand it to mean different things. Many papers have been written and no doubt will be written on this subject alone. However here we mean that we try to get away from overhaul lives fixed for a given engine type and move towards lives that differ between individual engines and their components, and that depend on their physical condition, history of use, and whether their failure would actually hazard the aircraft. In practice the rules are applied to varying degrees on RAF engines in service depending on the type of usage monitoring equipment fitted to the aircraft concerned.

As far as the turbine end of engines is concerned, our general rule is that because turbine discs are prone to low cycle fatigue (LCF) and because any failures might not be contained within the engine, lives based on engine usage are established. The approach for turbine blades is rather less straightforward and depends upon the likely failure mode. Creep and fatigue can probably be covered, but high cycle fatigue presents problems. In general, because failed blades would stay within the engine we tend to impose no in-service life and just let the blades fail.

3. ENGINE USE AND LIFE ESTIMATION. The subject of engine usage and life estimate is introduced at this stage because many people will be already aware of UK activities in this area from the detail in References 1 and 2. However a broad review is necessary because of its importance in relation to our other work.

Since our first tentative steps with an installation in a single Phantom in 1974, it has become increasingly apparent that total engine usage information was essential for two aspects; firstly to be fed back into aero-engine, and hence materials, design, development and testing. Secondly it provided for a more economic use, in-service, of the lives of engine discs obtained from theoretical and rig test sources.

In essence; the designer can now know how we actually use an engine as opposed to how we said we intended to. We can reduce the amount of unused safe life left on discs when we throw them away.

3.1 Engine Usage Monitoring System (EUMS) Mark 1. EUMS Mk 1 is our basic equipment. It is used primarily to produce a fleet exchange rate between low cycle fatigue cycles and flying hours or sorties. More than 15,000 engine ground and flight hours have been accumulated on it. Twelve different types of aircraft are fitted with the system and seven more, Tornado, Sea Harrier, VC 10 tanker, Chinook, Puma, Jetstream and Sea King, are planned for future fitment. Normally a sample of 4 to 6 aircraft of each type has EUMS fitted.

Variations in engines, pilot operation, and sortie pattern are taken account of using statistical techniques and are written into a computer programme used by the RAF.

The following table, also used in Reference 2, shows the safe cycles per hour, on average, which would be expected for various alternatives. In this case the figures happen to be for a compressor disc currently in service.

Monitoring Scheme	Safe average cycles per hour	
	LP	HP
None	3.26	1.73
By role and side of aircraft	2.56	1.39
By role, side and sortie pattern code	2.42	1.31
By LCF counter	2.27	1.26

As a measure of the success of the system, the LCF fleet exchange rate for the Adour in Jaguar was reduced, allowing time-expired Group A components to be reclaimed from quarantine at a cost saving of £1.3 million. In the same way EUMS data from the Pegasus engine in Harrier is being used for life extension. An average increase of 20% on turbine discs has been agreed.

The RAF Red Arrow display team, now operating British Aerospace Hawk aircraft, present a special problem. EUMS has shown that the outside aircraft in the formation can use about 13 times more LCF life than the formation leader on the same sortie. The need for recording on each Red Arrows aircraft will be covered later.

Not only does EUMS data provide the engine designer with a knowledge of throttle movements and therefore cyclic life consumption on aircraft in service but the data can be used to predict engine behaviour in future aircraft. In order to simulate the mission profile for projected aircraft, a EUMS fitted existing aircraft can deliberately reproduce the mission. Where the new mission profile is very complex, EUMS sorties can be flown on several types of existing aircraft and spliced together. This exercise is also very valuable in providing for the differences between peace-time and war-time operations at an early stage.

Currently the UK establishes disc life on 'life to first crack' principles. The RAF is very interested in any progress on the latest materials towards a Fracture Mechanics approach where cracks were permitted up to a certain size; although we are conscious of enormous practical difficulties of managing this in complete engines and ensuring its cost-effectiveness. However, to obtain the maximum benefit, it would not be sensible to have to use "nominal" mission criteria based on the worst flying profile. Therefore, engine useage monitoring on each individual aircraft appears to be a necessary step in this case.

As well as EUMS fitted to aircraft, a number are in operation on test beds. Here experience is gained and LCF/hot end life consumption rates can be quantified during the 150 hour Type Test. Moreover, the system can be used for more specific tests such as an assessment of the use of optical pyrometers to measure turbine temperatures directly on military engines. The results can then be read across to subsequent ones when the aircraft flies with the same modification.

The RAF is currently undertaking a trial using 12 Hawk aircraft fitted with EUMS primarily to explore the benefits of engine flight data recording and analysis using a Ground Processor based at the flying station. Detailed fleet management is the area most likely to show results. However, the volume of data will be large and special arrangements have been made to relate it to engine ground performance testing and the physical state of the engine found during overhaul. Much will be added to the engine designers knowledge of the Adour engine itself.

In-service experience with EUMS Mk 1 led to consideration of areas where its capability could be usefully increased and where extra features consistent with an overall engine life monitoring approach, could be added. The result was EUMS Mk 2.

3.2 <u>EUMS Mark 2.</u> Plessey Electronic Systems at Havant have developed EUMS Mk 2 to a UK Ministry specification. It is a microprocessor based system which includes a display suitable for either cockpit or equipment bay mounting giving parameter exceedance warnings and access to engine life data without having to use a ground replay equipment.

The system can accept 40 analogue parameters plus 16 spool speed inputs and 20 discrete ON/OFF signals. Up to 50 parameters can be checked for limit exceedances. LCF consumption is available for a number of critical components, both as a rate per flight and accumulated total. Data can be stored in either solid state memory or Quick Access Recorder.

EUMS Mk 2 is already flying in the Jaguar aircraft and is recording airframe fatigue as well as principal engine life parameters. The system has now been installed in the Harrier.

The overall management of EUMS Mk 2 and its data recovery follow closely the arrangements established for EUMS Mk 1.

3.3 <u>Low Cycle Fatigue Counter (LCFC).</u> As a partner to EUMS fitted to a sample number of aircraft, the LCFC provides the possibility of a monitor on each individual aircraft. The LCFC which is manufactured by Smiths Industries to a UK Ministry specification is a microprocessor based equipment which computes the LCF life consumed by 4 specific features in the engine rotating components. It notes rpm excursions and converts them into reference stress cycles. The RAF evaluation has taken place on aircraft already fitted with EUMS so that the algorithms used in the LCFC can be cross-checked.

Most of our results have been obtained on the Adour engine in Hawk and Jaguar. Its primary use so far has been on the Red Arrow display team aircraft where, as mentioned earlier, wide variations in LCF consumption can occur.

At the moment the counter neglects thermal effects and the simple LCFC is clearly restricted where thermal gradients or thermal fatigue are significant. As it happens, it is safe to ignore thermal effects on a large number of RAF engines but this will be less possible in future. Three further approaches are being worked on. Firstly, the LCFC is being developed to take account of turbine blade thermal fatigue and creep. Secondly, it may be possible to adjust the value of Ultimate Tensile Strength, used in the LCFC, to account for changes in material temperatures. Finally, EUMS Mk 2 could be fitted to each aircraft allowing more parameters to be measured.

3.4 Hot-end Monitoring. For the turbine area, in particular, the service engineer needs an on-board hot end monitor that he can trust. Backing by a proven analytical basis is essential. The design process, including test programmes is fairly well established. The EUMS programme will be used to confirm life predictions. Rolls Royce and NGTE have developed creep and thermal fatigue algorithms which will be validated by phased removal of components and metallurgical examination.

By ensuring that measurements of compressor outlet temperature (turbine blade cooling air), turbine inlet temperature and shaft speed are available, the spare capacity built into the LCFC could be used to assess creep and thermal fatigue. This facility will be available soon. However, a programme of EUMS correlation will be essential before the LCFC could become executive for these quantities. Even then both hot end equations have limitations because the failure mechanisms are not completely understood. However, a period of steady investigation should improve this situation.

3.5 Modular Engines. Before turning to the practical activities of inspection and repair, some broader mechanical constraints should be considered because they affect how and when these activities can be done.

The RAF is firmly committed to modular engines because of the large cost savings from the reduced buy of complete spare engines. Moreover, engines can be recovered for use more quickly in the field. The newer engines have up to 17 modules. Most inspection and repair work is now done at flying stations with engines and modules only returning to an RAF Maintenance Unit or Industry when essential. This balance is the complete reverse of that employed in past years and, of course, the RAF still has large numbers of 'traditional' engines in use.

The very flexibility of the modular engine and our in-service overhaul system also seriously compounds the management problem. For instance, we are progressing from the relatively simple task of managing a fleet of engines with the same life, through the stage with our Phantom and Nimrod engines which each have one or two different module lives to the multi-modular engine like the RB 199 example on a station operating 50 twin-engined aircraft; instead of 100 individual engine lives, we have 2,900!

It is for this reason that we have been forced to install a local station level ADP system based on the Texas Instruments 990/10 minicomputer.

3.6 Inspection. In an ideal world the maintenance engineer would like to be able to inspect all parts of the engine without dismantling anything and without removing it from the aircraft. Moreover, all defects should be so obvious that no inspector would miss them even at the end of a tiring work shift. Real life does not allow such luxuries. To be fair, facilities for inspection have probably never been better. On the other hand there is more to look at; the number of blades is a prime example. Also the tendency towards more complicated fabrication reduces access to the areas to be inspected.

3.6.1 Inspection of the Assembled Engine. Here visual inspection still predominates. These days the engine is well provided with access ports for borescopes. Even the problem of preventing access port plugs in the turbine area from seizing up seems to have been solved.

It has been clear for some time that the strain of lengthy unaided borescope inspections greatly increases the risk that an inspector will miss a defect. Closed circuit television (CCTV) is an essential step both to ease the strain on the inspectors and also to allow a group to view and discuss a defect together and, if necessary, to take a photograph.

Nevertheless black and white television can also be misleading and suggest spurious defects which turn out to be merely discolorations of the material. It is most likely that we will soon turn to more widespread use of colour television to counteract this problem.

X-Ray techniques are sometimes used; either by X-Ray generation from the outside or by the introduction of a radio-active isotope into the internal parts of the engine. In general, the use of X-Rays is far from straightforward because the many layers of fabrication cause a very confused and complicated X-Ray picture. Even so the technique has been valuable for some very specific applications.

3.6.2 Inspection of the Dismantled Engine. The most recent changes in this area for the RAF have been introduced less by new techniques and more by where, in organisational terms, the inspection is done. As noted earlier the station level engineers now see more of the engine internals than at any time in the past. However, most module inspections at stations are limited to visual techniques; both aided and unaided. Visiting non-destructive testing (NDT) teams supplement this. The RAF has developed, over the years, a fair capability for NDT techniques; spurred on by airframe structural needs. For engines and modules which return to Maintenance Unit level, the full range of NDT devices is available for use and most of our present effort is directed towards semi-automation of these techniques to increase productivity.

3.7 Repair. On RAF flying stations very little repair of high temperature materials is undertaken except by replacement of complete components such as turbine blades. However, removal of modules on stations exposes other modules to view. As a result, cracks in, say, the combustion chamber become known about. Whereas in the past if the engine was performing satisfactorily, they remained invisible until the engine was finally dismantled. This new situation has increased the flow of questions to the materials designer about acceptable crack or damage limits. Here it is important that the advice given is sensible and that defects that we can live with are clearly identified. Wherever possible, the over-safe and expensive route of trying to repair everything should be avoided.

Our modular policy introduces a further risk that, because the engine internals are exposed more often, they will even with our best control be exposed to minor and sometimes unnoticed surface damage. This is a fact of life for the future and the materials designer should bear it in mind. Although how he could possibly cater for such a random event is difficult to imagine.

Corrosion and erosion are always with us. However, they are areas where military operations can still catch out the maintenance engineer and indirectly the materials designer. Sometimes our aircraft are eventually used in roles and environments never expected when they were first specified. A classic case is that of the Harrier used in Belize, where severe turbine disc corrosion/erosion was initiated by the unusual environment.

At Maintenance Unit level the RAF share full engine repair with Industry. The cost of repair equipment required for modern materials limits its use to a few locations. Electron beam welding equipment, plasma spray and extensive heat treatment plant appear to be an inevitable price to pay for materials now coming into use.

3.8 Costs. Although the need to contain costs has driven most of what has been said already, it is notoriously difficult to pin them down. In general the value of the aero-engines owned by the RAF is about £1,400 million and rising rapidly with the introduction of the Rolls Royce RB 199. What is more, by the time an engine finally leaves the Service, supporting it will have cost us at least 3 and probably more, times its purchase price, and that excludes the fuel it has burned.

It is therefore usually easy to justify work on more precise lifing and increasing the capability for inspection and repair.

4. CONCLUSIONS. Cost constraints and the need for increased engine availability has provoked a much more flexible approach to the estimation of the in-service life of engine parts. Therefore the UK will continue to operate and expand an engine useage monitoring programme, exploiting the strengths of the various equipments available, in order to refine component lives and move further towards an On Condition Maintenance Policy. At the same time the flow of data back to the designer should fill earlier gaps in the information about military engine use. Good analytical foundations for predicted failure modes are needed early to identify the parameters to be measured and the algorithms to be used.

Techniques of inspection and repair in-service will, of necessity, evolve with future changes in materials and this need should not be overlooked in the search for materials bringing performance gains.

Although much of what has been considered applies equally outside of the high temperature materials area, this area does involve the shortest in-service lives and greatest difficulties of inspection and repair. Looked at positively, improvements in these materials therefore bring the greatest potential for general cost and operating benefits.

ACKNOWLEDGEMENTS

The author wishes to acknowledge the co-operation of associates in Industry and also to his colleagues in the Ministry of Defence for assistance in the preparation of this paper.

The views expressed in the paper are those of the author and do not necessarily reflect Ministry of Defence policy.

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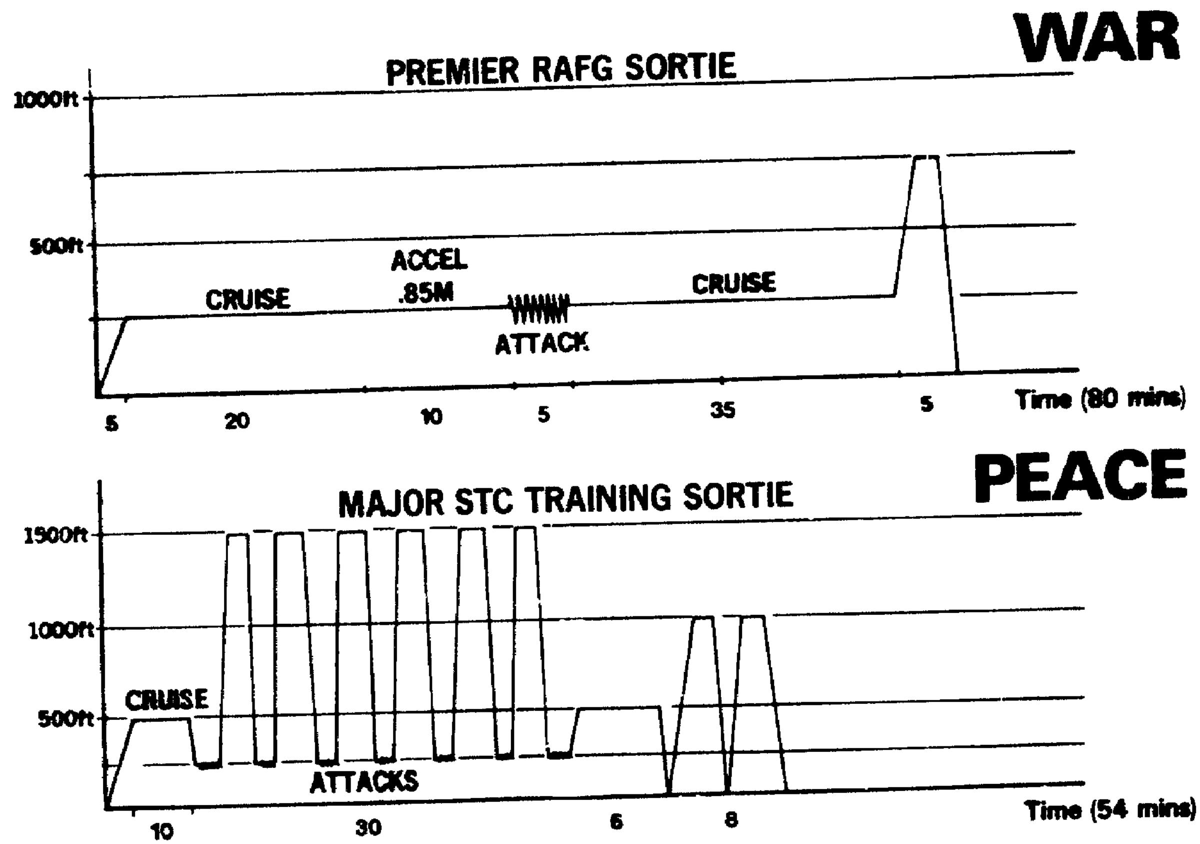
WAR
PREMIER RAFG SORTIE
1000ft
500ft
CRUISE
ACCEL
.85M
ATTACK
CRUISE
5
20
10
5
35
5
Time (80 mins)
PEACE
MAJOR STC TRAINING SORTIE
1500ft
1000ft
500ft
CRUISE
ATTACKS
10
30
6
8
Time (54 mins)

FIG 1

ENGINE DEPOT MAINTENANCE REPAIR TECHNOLOGY

by

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The scope and mission of the two USAF engine Air Logistics Centers are described. The various processes and organizational structure to identify repair technology requirements are discussed. Approaches to transition and implementation of new technology into a repair depot environment are described. Specific examples of technology developments described are: braze repair, laser metrology, electrophoretic, coatings, sputtered MCrAlY overlay coating and inlet guide vane vibration damping.

INTRODUCTION:

The Air Force Logistic Command (AFLC) is the command which is responsible for the overhaul, repair, spare parts and supplies necessary to keep Air Force weapon systems combat ready. The responsibilities include aircraft, missiles and support equipment. As can be imagined, this is not a simple or small task. If the resources of AFLC were compared to those of U.S. corporations, AFLC would be ranked 11th in the Fortune magazine list of the top 500 U.S. companies. In order to satisfy its mission, AFLC functions on both a contractual and organic or in-house basis. Five Air Logistics Centers (ALCs) have been established to provide the support to specific weapon systems or components. The primary workload assignments for each ALC are shown in Figure 1.

The organic repair functions of AFLC represent a significant percentage of AFLC resources. The value of AFLC maintenance facilities and equipment exceeds $1.3 billion with covered floor space in excess of 16 million square feet. Thirty-five thousand people are employed for organic maintenance. In terms of aircraft and engines, these facilities perform maintenance and repair of about 50% of the aircraft and over 80% of engines overhauled in Fiscal Year (FY) 1980.

Overhaul and repair of engines are accomplished at two Air Logistics Centers (ALCs), Oklahoma City ALC and San Antonio ALC. During FY 80 approximately 4,600 engines or engine modules, in the case of F-100, were overhauled by AFLC. Over 3,900 of these were accomplished on-site at these ALCs. Engine repair resources during FY 80 totaled $210 million. The distribution of the engine workload for FY 80, is shown in Figure 2.

For many years, Logistics Command was not viewed by the R and D community as a viable customer for new technology. The reasons for this are many and varied. For example, the cost of ownership of a weapon system was predominantly in development and acquisition as opposed to operation and support, the logistics community was not requesting support, from the R and D community, and many of the problems encountered by AFLC did not lend themselves to waiting for long term solutions implied by R and D, and many laboratory personnel were not interested in applying themselves toward solving what they considered to be mundane problems.

The close association between the Air Force Logistics Command and the engineering community of the Air Force Systems Command has resulted in the identification of various opportunities for the translation of advanced materials and process technologies to the overhaul floor. These advanced technologies offer the potential for reduced costs during engine overhaul. Several examples of programs which have or are being conducted to enhance the repair of fan and turbine blades will be described.

BRAZE REPAIR FOR TURBINE AIRFOILS:

Advanced turbine blades and vanes require the use of sophisticated air cooling techniques, costly nickel and cobalt base alloys, and extensive surface protective coatings. Because their operating environments cause various types of degeneration which ultimately lead to their removal and replacement, cost effectiveness of repair versus replacement must be considered in terms of overall life cycle management. The Air Force Wright Aeronautical Laboratories - Materials Laboratory (AFWAL/ML) conducted a Manufacturing Technology (MANTECH) program with General Electric Corporation; (1) to establish cost effective repair techniques for conventionally cast airfoils. The objectives of the program were: 1) to select repair processes and airfoil types with generic application to ALC repair requirements, 2) to transition advanced process to manufacturing technology, and 3) to verify the repair procedure.

Components selected for repair were the TF39 first and second stage turbine vanes and the first stages turbine blade. These are shown in Figure 3. Repairs performed on the Stage 1 HPT vanes incorporated cracked areas of the leading edge, trailing edge (concave surfaces) aft of the cooling holes and various inner and outer platform locations. In order to make a braze repair, the Fluoride-Ion Cleaning Process was developed in order to remove the oxide from the surface of the crack. A schematic of the cleaning process and its effectiveness in oxide removal are shown in Figure 4. Activated Diffusion Healing (ADH), which is a hybrid brazing process, was used to repair the cracks. The results of a typical ADH repair of a vane are shown in Figure 5. Representative components of each of the Stage 1 and Stage 2 vanes were subjected to 1,000 mission profile cycles in TF39 engine test vehicles, and inspected, and then evaluated to determine the degree of success for each repair. Repair feasibility for both components was satisfactorily demonstrated. The economics of the process for turbine stator components demonstrates the cost effectiveness of the repair concept in lieu of new parts replacement for total engine life cycle management. Additionally, the advanced techniques developed in this program using Fluoride-Ion cleaning and ADH extend the capability and cost savings afforded in turbine airfoil repair.

METROLOGY OF COMPLEX AIRFOILS:

Measurement of dimensional characteristics of airfoil parts in both production and overhaul is primarily a manual operation. These measurements employ a variety of gage types. Gages vary from the very expensive optical comparator (Figure 6) to the inexpensive dial gage (Figure 7). Some gages are simple, others quite elaborate. All are subject to wear and require their own individual calibration procedures. All gages require care in handling and some of the more delicate ones require frequent repair. The amount of training, skill and judgment necessary to use these gages varies considerably with the type of gage employed. Regardless of the dimensional gage used, alert and conscientious inspection is required by trained inspectors.

Despite the diversity of gages and the heavy reliance on manual inspection, dimensional inspection today h satisfactory accuracy and provides adequate reliability. It is clear, however, that an automatic non-contacting inspection gage capable of measuring most dimensional characteristics of interest would: greatly simplify inspection operations; be cost effective; consolidate a number of gages into one; minimize wear through the non-contacting feature; and improve overall inspection reliability by minimizing human involvement. The need for such a system is further emphasized if future needs are considered, e.g., the potential for increased airfoil parameter measurements and tighter tolerances that could result from the need for improved engine performance reliability. An automated system would address to future needs as well as provide many improvements for inspection of today's hardware.

The features desired in an automatic gage establish the basic area for focus during its development. Rapid inspection speed is important for the automatic gage to be cost-effective. Accuracy is necessary to meet design requirements, and repeatability is important to provide consistent results. Non-contacting probes preclude scratching of the part and probe wear problems, and provide greater accessibility to part surfaces. Universality of application is desirable to minimize the need for many gages and their associated costs. Important ingredients of universality are ease of fixturing and flexibility of software to accommodate a variety of parts and inspection requirements. Automatic documentation must provide necessary information for proper part disposition in an easily understood format. Finally, the automatic gage must be compatible with the production and overhaul environment. All of these features must be considered and included, to the extent possible, in the design and construction of an automatic gage.

The AFWAL/ML conducted a MANTECH program with General Electric Corporation (2) and Diffracto, Ltd. of Windsor, Ontario, Canada to develop and establish and validate a semi-automated, non-contacting gaging system to make accurate measurements. This program was initiated to improve this technology area by establishing a non-contacting, computer interfacial gage system that provides the capability of measuring many dimensional characteristics (e.g. airfoil contour, length, width, twist, lean, etc.) with one instrument while maintaining high accuracy, repeatability at reasonable throughput rates.

The gage system was to be a demonstrator semi-automatic laser gage system capable of measuring representative dimensional characteristics of single turbine blades and single turbine vanes within a part size envelope approximately 12 inches long, 4 inches wide and 4 inches deep. The demonstrator system was to be capable of measuring ten representative dimensional characteristics of a test part. The test part for this program was

the TF39, stage 2 trailing high pressure turbine blade.

The demonstrator laser gaging system is a semi-automatic four axis system capable of measuring the ten representative dimensional characteristics of the test part. The demonstrator system (Figure 8) consists of four primary modules: a) a laser sensor module; b) a part manipulation module; c) an output module and a computer module. The laser sensor and part manipulation modules comprising the demonstrator gage which interfaces with the computer module is shown in Figure 9.

The program essentially met the design criteria that the demonstrator system have an accuracy plus repeatability of 15% (i.e. +90 microinches) of the tightest tolerance to be measured at the 95% confidence level. The overall system met this requirement over an estimated 80% of the part size envelope and within 11% of the goal over the remaining 20% (i.e. the outer radial periphery). The major objectives of the program were achieved and a firm technological foundation has been established for a cost effective, reliable and flexible follow-on production model for airfoil part dimensional inspection.

ALUMINIDE PRODUCTION COATING PROCESS:

Various aluminide type coatings are used to protect turbine hardware in Air Force engines. While these coatings are similar in composition and performance, their application methods vary so that coating new and used parts may require separate installations unique to the processes. For example, the various steps in a pack process are shown in Figure 10. Cost would be reduced and logistics simplified if a single coating process were available at the Air Logistic Centers (ALC).

The AFWAL/ML conducted a MANTECH program with Detroit Diesel Allison (DDA) Division of General Motors Corporation to provide a single process coating capabilty at a selected Air Logistic center. The Allison Electrophoretic Process (AEP) is to be employed for coating nine different alloys (nineteen components) in the Air Force engine inventory. It is expected that through the use of AEP the Air Logistics Centers could satisfy all their refurbishing needs with two basic bath chemistries - AEP 32 (Al-Cr-Mn) for nickel base and AEP 100 (Al-CrAl) for cobalt base superalloys - both being applied in a common installation. The two coatings proposed are licensed for use by the U.S. Government per contracts DAAJ-69-C-0412 and F33647-72-C-0007. Laboratory and engine testing are to be used to evaluate the suitability of AEP as a single coating process for Air Force use.

The AEP Coating Process is based on the principle that migration is observed when an electrical potential is applied to two electrodes immersed in a dispersion of charged particles. An electrophoretic cell is shown in Figure 11. When the dispersion is properly formulated, the suspended particles will deposit upon one of the electrodes. Electrophoretic deposition differs from electroplating in that particles of any composition rather than ions are deposited. As compared with physical vapor deposition, electroplating spraying or dipping, the electrophoretic procedure is particularly suited to applying coatings to non-uniform geometries, because, as the coating deposits on external areas and edges they become insulated and the effective deposition shifts to uncoated areas. The AEP procedures utilize fine particles dispersed in a low viscosity polar liquid. Emperically determined coating weights are electrophoretically deposited upon turbine components. The coated articles are then subjected to thermal treatments in suitable environments (hydrogen and argon have been used successfully) for times necessary to produce substrate-coatings with the desired structure.

Facilities required for the AEP process are relatively simple and cost substantially less than for other conventional aluminide processes. A production facility has been operational at DDA for more than five years applying AEP 32 coating to Model TF 41 1st Stage Turbine Blades. Although this facility occupies only 250 square feet of floor space, it has a capacity for applying "green coating" to 1200 blades per eight hour shift. A typical AEP coating facility and process steps are shown in Figure 12 and 13.

An AEP aluminide coating can be tailored to meet a variety of specific turbine components requirements. The AEP 32 (Al-Cr-Mn) composition was developed to overcome thermal cracking problems encountered with the DDA pack cementation coating (Alpak) and at the same time hot corrosion resistance was improved. The AEP 32 coating has shown excellent performance and applicability to a variety of superalloy substrates in previous AFWAL/ML sponsored programs (5,6). The resistance to hot corrosion, low and high cycle fatigue and stress rupture of various nickel base alloys coated with AEP 32 and alternative coatings are shown in Figures 14-17. Micrographs of the PWA 73, Alpak and AEP 32 coatings on two nickel base alloys are shown in Figures 18 and 19. Similarly AEP 100 (Al-CrAl) was developed to overcome Alpak coating thermal cracking and spalling on X40 and Mar-M509 turbine vanes, and is the prime candidate for improving turbine vane assemblies in Model T56 and DDA Industrial Gas Turbine Engines.

The AEP 32 (Al-Cr-Mn) and AEP 100 (Al-CrAl) coatings have been successfully applied to a variety of turbine engine components at DDA for more than seven (7) years. This includes both solid and air cooled turbine blades, turbine nozzle vane assemblies ranging from single to as many as six integral airfoils, integrally cast turbine wheel and blade assemblies and full-round integrally cast nozzle vane assemblies.

Uniformity of finished coatings has been of demonstrated excellence; e.g. five years of production for the Model TF 41 1st Stage Turbine Blades with a specified coating thickness of 1.5 to 3.0 mils on gas path surfaces, airfoil, shroud, platform, and stalk, has shown a coating thickness uniformity over the coated surfaces of individual parts in

the range of 0.5 mil variation with no difficulty in meeting the overall specified thickness range. Similar success has been experienced with integral multiple vane and wheel assemblies.

As a part of the current MANTECH program, DDA is to install an AEP Coating Unit at the San Antonio Air Logistic Center (SAALC) to coat blades. The SAALC plans to coat initially the T56 blades and later the TF39 blades. Similar AEP coating facilities are planned for the Oklahoma Air Logistic Center that overhauls the TF41.

MCrAlY OVERLAY COATING PROCESS:

A class of coatings (the MCrAlY overlay type) has been developed that significantly increases the durability of turbine airfoils over presently used diffusion aluminide type coatings. The MCrAlY overlay coating is deposited on the airfoil surface and provides, independent of the base alloy composition, increased resistance to oxidation, hot corrosion, and fatigue. Because of the performance of these coatings and the fact that they can be readily tailored in composition and microstructure, they are finding increased usage in advanced military gas turbine engines. Such overlay coatings as NiCrAlY, CoCrAlY, and NiCoCrAlY, and improved compositions being developed, will provide the next generation of coatings for advanced military gas turbine engines.

Increased durability and higher turbine airfoil operating temperatures are primary requisites for these future engines. Both of these needs translate to an absolute requirement that better-than-present protective coatings be integrated into the design and specification of the turbine blades and vanes. To meet the need, extensive coating development efforts are being pursued. Many, if not most, of the efforts to develop improved overlay coatings are limited by the restricted capabilities of processing methods to apply tailored coating compositions and structures on an experimental and ultimately, on a production or overhaul basis. With the current electron beam vapor deposition (EB-PVD) process (used to apply today's MCrAlY coatings) it is difficult to deposit complex compositions containing low vapor pressure elements as Ta or Hf.

The AFWAL/ML conducted a MANTECH program with Pratt and Whitney Aircraft to establish and optimize an advanced overlay coating process which is suitable for application of MCrAlY type overlay coatings to high-pressure turbine blade and vane components during manufacturing and for recoating during overhaul for Air Force advanced gas turbine engines. The process is to be adaptable to both nickel and cobalt base superalloys with only minor process variations, such as adjustments or modifications of coating composition and changes in coating morphology. An additional objective of the program is to provide a technique with equivalent or superior performance to commercially available vapor depositions coating processes with definable cost savings relative to advanced coating requirements.

Innovations in plasma-spray technology have resulted in significant improvements in the metallurgical quality of plasma-sprayed overlay coatings. However, laboratory and engine tests of NiCoCrAlY coatings within P&WA indicate that the best current technology plasma-spray performance is 60 or 70% that of equivalent electron beam coatings at temperatures and conditions defined for advanced gas turbine engines; therefore, sputtering was selected for further development and scale-up (7).

The sputter coating process is a vacuum coating technique in which inert gas ions, typically argon, are accelerated from a plasma into the source, or target, material to be deposited. On bombarding the target, the ions eject atoms and/or molecules from a thin layer on the target surface to create a coating flux from the target surface. The part to be coated is held in the flux, and the ejected or sputtered material collected and recombined to form a solid coating representative of the target material composition. A schematic of the sputtering process is shown in Figure 2. The process shown is termed triode sputtering, due to the use of a thermionic emitter to increase the degree of ionization in the plasma. The cathode sputtering system is shown in Figure 21.

A post-hollow target (cathode) geometry was used in this program for process optimization. A target configuration having two concentric, cylindrical targets. Figure 20, was selected as the best compromise of part volume-per-target area and deposition efficiency. With the parts to be coated, held, and rotated in the annulus between the two targets, coating material arrives at the part from both targets. Most of the coating flux material that misses the part deposits on the opposite target and is subsequently resputtered, producing a high material utilization efficiency.

A typical microstructure of MC-AlY sputtered coating deposited at low substrate temperature is shown in Figure 22. The apperance of the sputtered NiCoCrAlY coated blade after engine operations is shown in Figure 23. Because of the fundamental principles of the sputtering process, it has the flexibility to deposit coating composition with exceptional structural quality combined with the relative ease with which the process can be automated, reproduced and controlled, it makes sputtering an attractive production and overhaul coating technique. Some of the inherent advantages of sputtering are shown in Figure 24. This process will soon be working its way from manufacturing technology to the overhaul floor.

VIBRATION DAMPING:

The inlet guide vanes of the TF30 engine for the F-111 Aircraft were experiencing premature failure in as little as fifty hours of operation. The analysis of the failures

revealed that the vanes were cracking due to fatigue as a result of the high stresses which developed at certain engine speeds. These excessive bending stresses in the inlet guide vanes were caused by the interaction of the incoming air with the inlet duct and the compressor. The initial solution required a redesign of the inlet duct and a retrofit of the new inlet duct.

This problem was analyzed by Dr. Jack Henderson of the AFWAL/ML. He developed a multiple layer energy dissipating adhesive and aluminum constraining layer which was adhesively bonded to the outside surface of the inlet guide vane. The selection of the adhesive, number of layers and surf. . treatment was complex because of the anti-icing requirement. A photograph of the damping wrap adhesively bonded to the inlet guide vane is shown in Figure 25. The effectiveness of the damping treatment can be seen in Figure 26. After engine and flight testing, the damping treatment was approved for application. This damping treatment is being installed on all TF30 engine inlet guide vanes during overhaul and is applied on new engines. Similar type damping treatments are being applied to the TF41 and TF33 engines. The success of this work has provided impetus for the development of high temperature polymeric damping materials and vitreous enamels. The application of damping treatment has demonstrated that this technology offers a solution to complex problems and an alternative to costly redesign.

CONCLUDING REMARKS:

In the past few years, considerable effort has been directed at reducing the cost-of-ownership of turbine engines by translating various advanced materials and process technologies to the overhaul floor. These technologies permit the repair of previously salvaged parts, enable parts to be inspected more reliably and more quickly, reduce the number of different coating facilities required and extend the life of overhauled parts. The various examples presented are just a sample of some of the efforts underway to address the problem of engine overhaul. Many opportunities exist in the engine overhaul environment for reduction of operation and support costs by the application of advanced materials and process technologies.

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AIR FORCE LOGISTICS COMMAND (AFLC)
(MAJOR TRC ASSIGNMENTS)

OKLAHOMA CITY ALC	OGDEN ALC	SACRAMENTO ALC	SAN ANTONIO ALC	WARNER ROBINS	AGMC
B-52G	F-4	A-10	B-52H	F-15	Inertial Msmt Units/Platforms Nav. Inst.
C-135	F-16	F-111	C-5A	C-141	Displacement Gyroscopes
E-3A		F-106		C-130	
A-7D	Weapons	CT-39	Electronic AGE		
	Airmunitions	FB-111	Electrical/ Mech. AGE	Airborne/ Electronics	
Oxygen Components	Missile Comp.	F-105	Nuclear/Com ponents	Life Support Sys	
	Landing Gear			Propellors	
Engine (J57, J75, TF30, TF33, TF41, F101) (Except RAM Air Turb)	Photo Equip	Electrical Components	Engines (J79, T56, TF39, F-100)		
	Training & Simulation Equip	Ground Elec (CEM)		Gyroscopes (except dis- placement)	
Pneumatic Accessories	Navigation Inst. (ex- cept IMUs/ Platforms)	Flight Con- trol Inst.			
Constant Speed Drive	Pneumatic Access. RAM Air Tur- bine	Hydraulics/ Fluid Driven Accessories			

FIGURE 1. AFLC MAJOR TECHNICAL REPAIR CENTER (TRC) ASSIGNMENTS

Figure 2. FY 80 ENGINE PRODUCTION (UNITS)

TMS	ORGANIC	CONTRACT	INTERSERVICE	TOTAL
J-57	392	203		595
J-75	74			74
TF-30	438			438
TF-33	136			136
TF-41	257			257
T-58			30	30
T-64			12	12
T-56	364	4		368
J-79	682			682
TF-39	27			27
F-100-100	97	23		120
F-100 Modules	1,293			1,293
G/T-400			24	24
J-60		42		42
J-65			12	12
J-69		105		105
J-85		126		126
T-53			2	2
T-76			37	37
TF-34			21	21
IGSO54		4		4
IO360D		87		87
O300-D		4		4
R1830		1		1
R2000		41		41
R2800		140		140
TOTAL	3,760	780	138	4,678

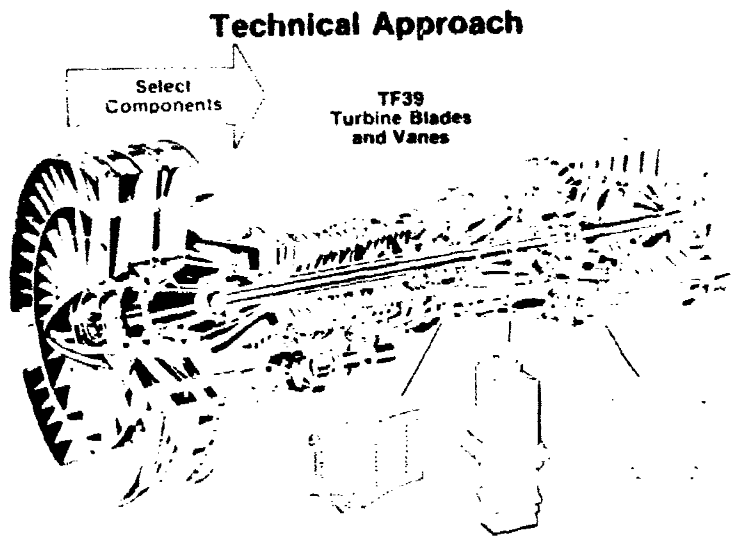


FIGURE 3. CROSS-SECTION OF TF39 ENGINE

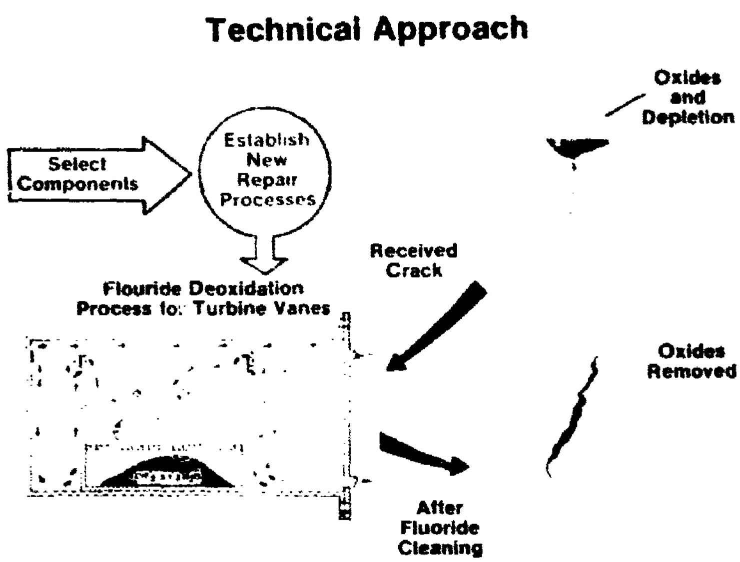


FIGURE 4. FLUORIDE CLEANING OF OXIDE BEFORE BRAZING

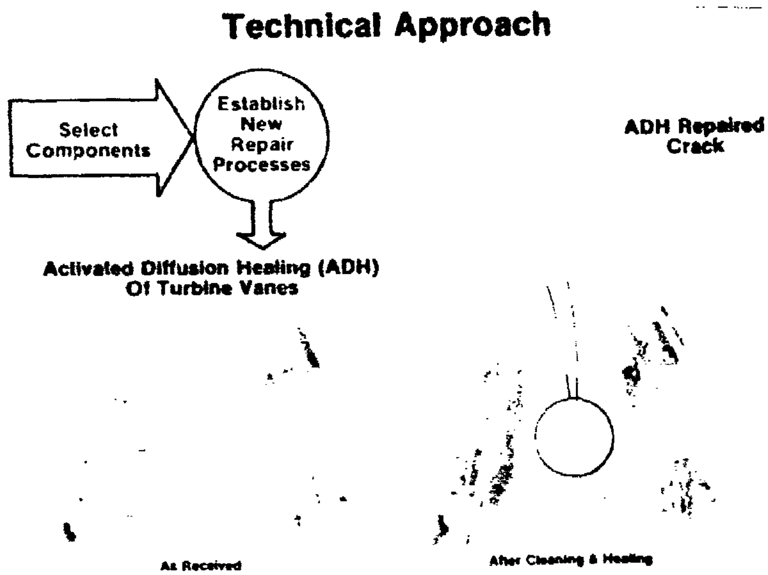


FIGURE 5. ACTIVATED DIFFUSION HEALING OF TURBINE VANES

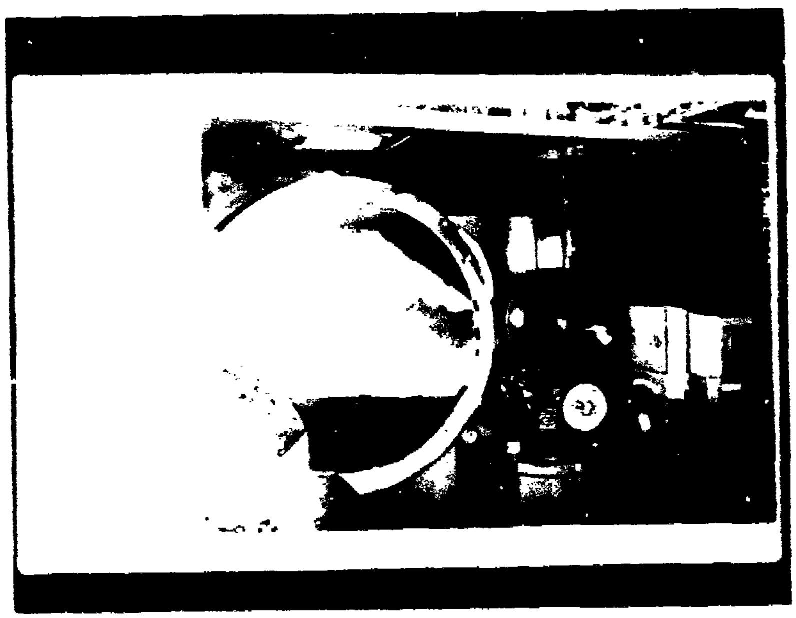


FIGURE 6. OPTICAL COMPARATOR GAGE USED TO MEASURE TURBINE BLADES

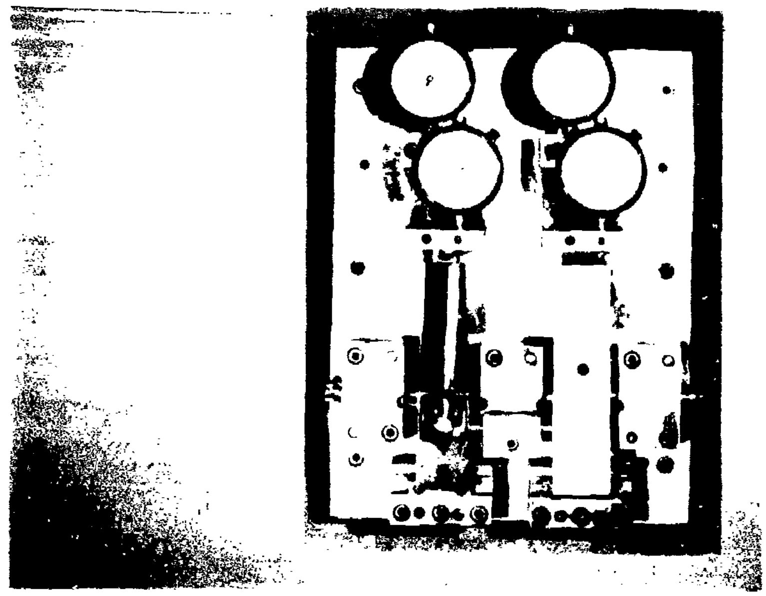


FIGURE 7. DIAL INDICATOR GAGE USED TO MEASURE TURBINE BLADES

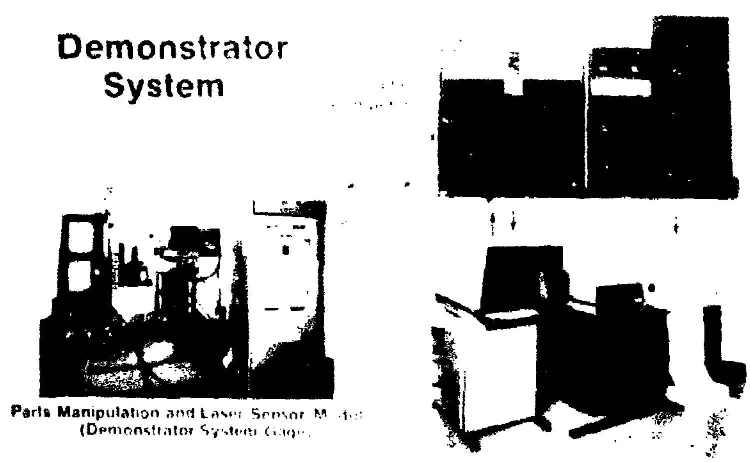
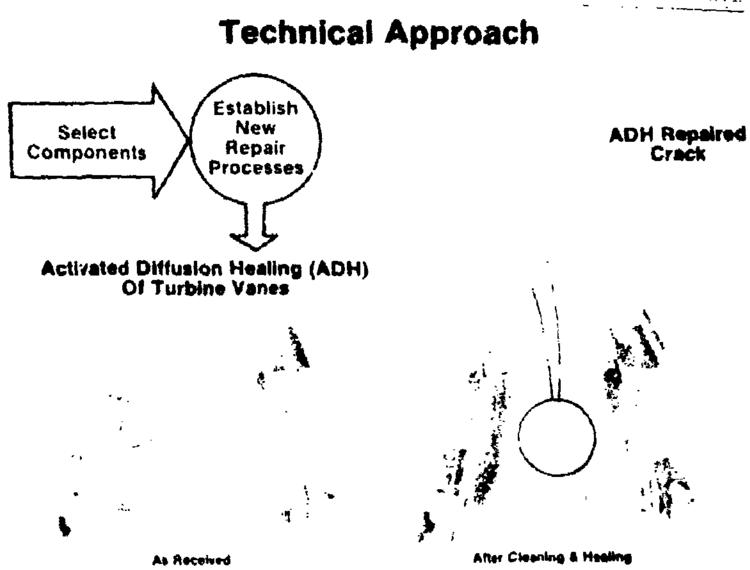


FIGURE 8. LASER METROLOGY DEMONSTRATION SYSTEM



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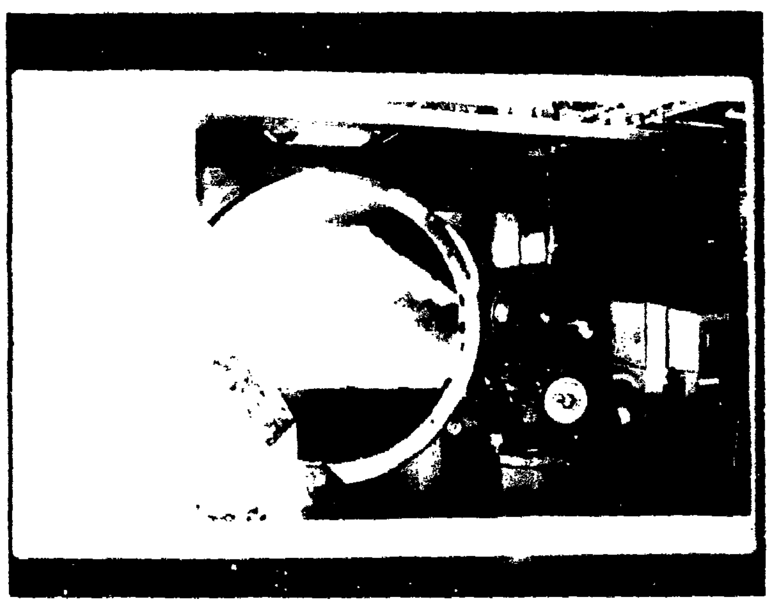

FIGURE 6. OPTICAL COMPARATOR GAGE USED TO MEASURE TURBINE BLADES

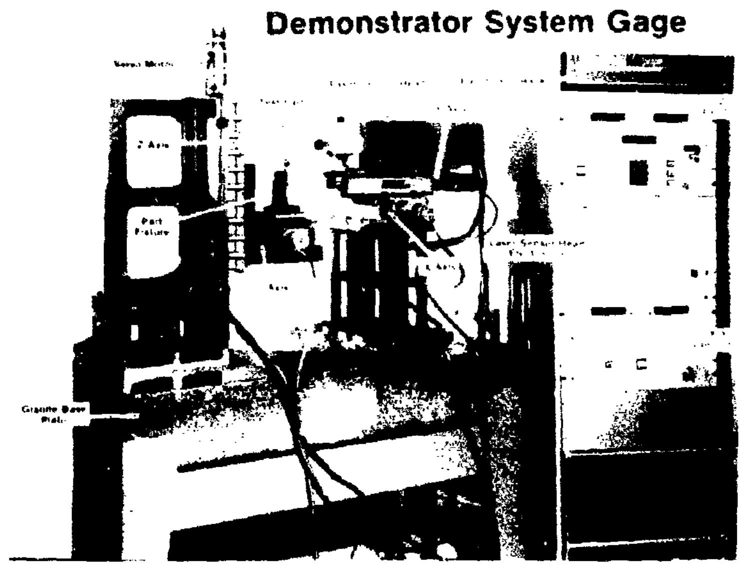

FIGURE 9. LASER METROLOGY SYSTEM GAGE

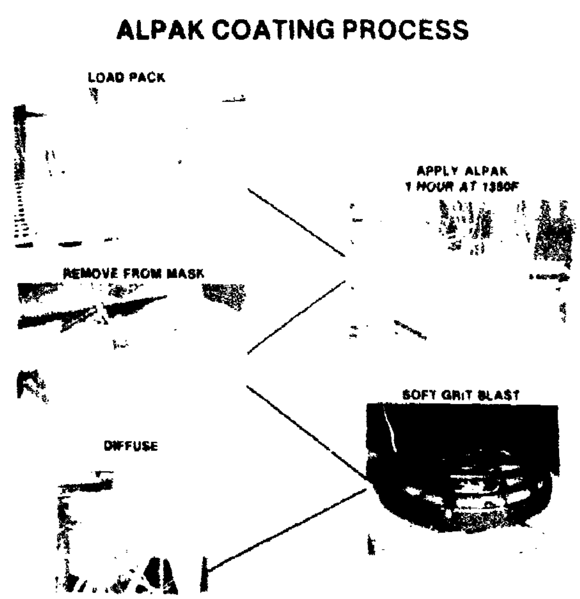

FIGURE 10. ALPAK COATING PROCESS

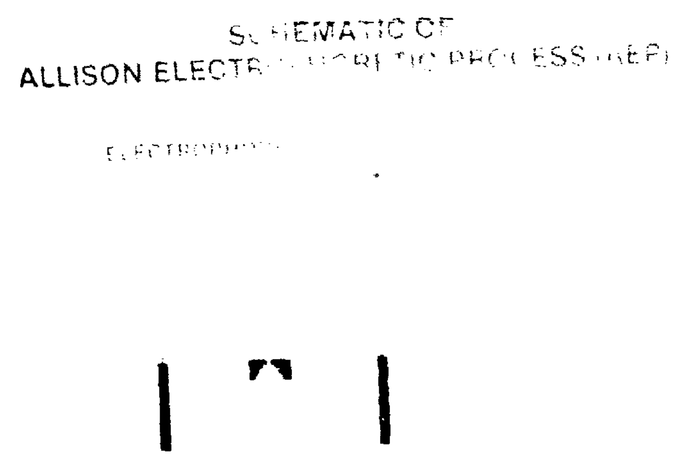

FIGURE 11. DIAGRAM OF ELECTROPHORETIC CELL

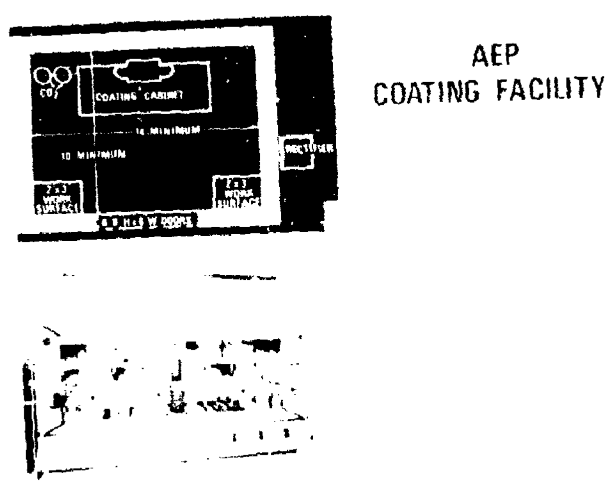

FIGURE 12. TYPICAL ELECTROPHORETIC COATING FACILITY

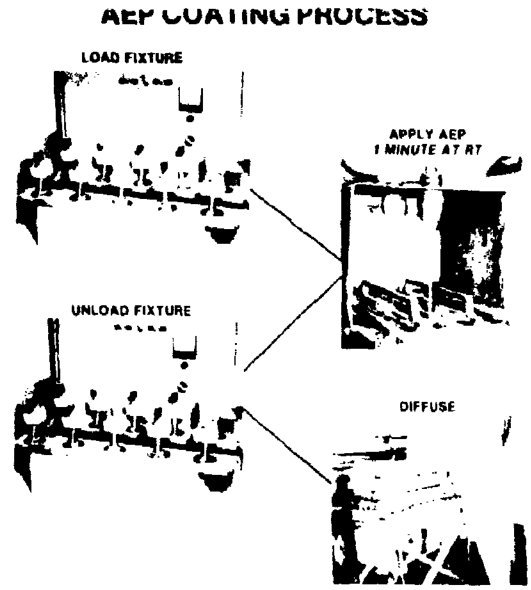

FIGURE 13. ELECTROPHORETIC COATING PROCESS

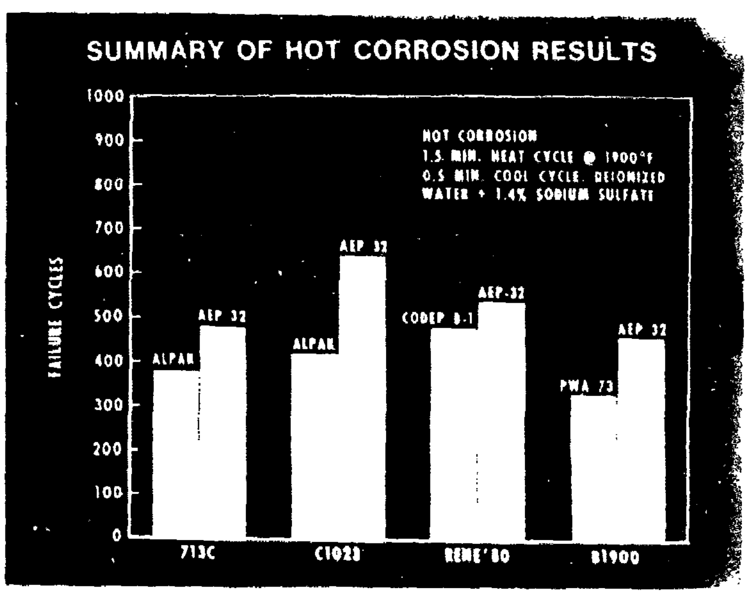

FIGURE 14. COMPARISON OF HOT CORROSION RESISTANCE OF AEP-32 COATED NICKEL-BASE ALLOYS

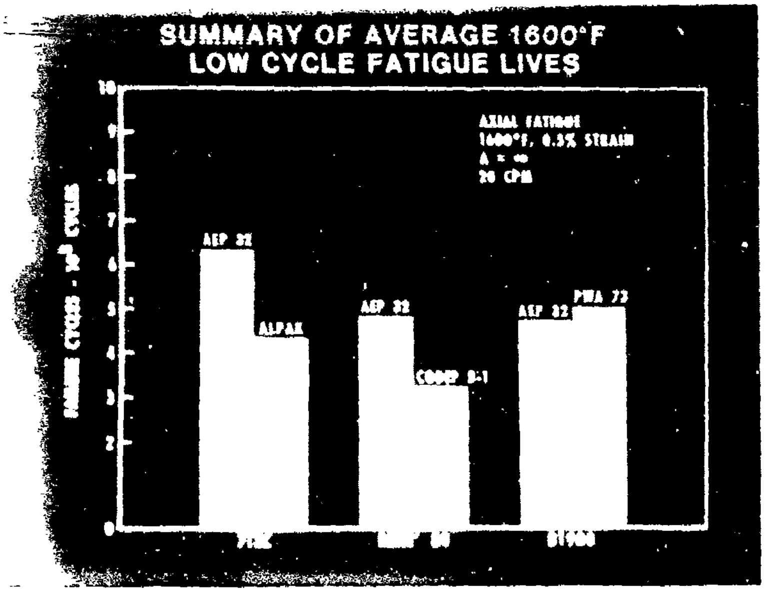

FIGURE 15. COMPARISON OF [illegible]

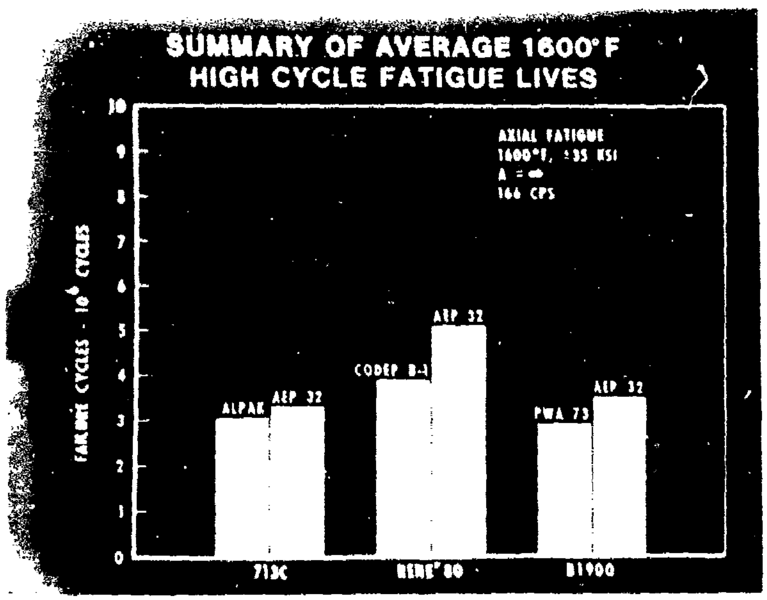

FIGURE 16. COMPARISON OF HCF LIFE OF AEP-32 COATED NICKEL-BASE ALLOYS

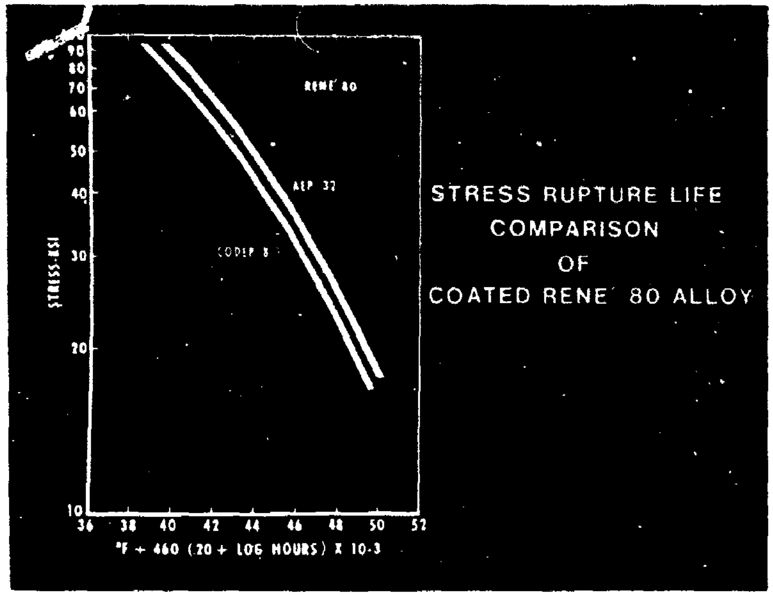

FIGURE 17. COMPARISON OF STRESS RUPTURE LIFE OF CODEP B-1 AND AEP-32 COATED RENE' 80

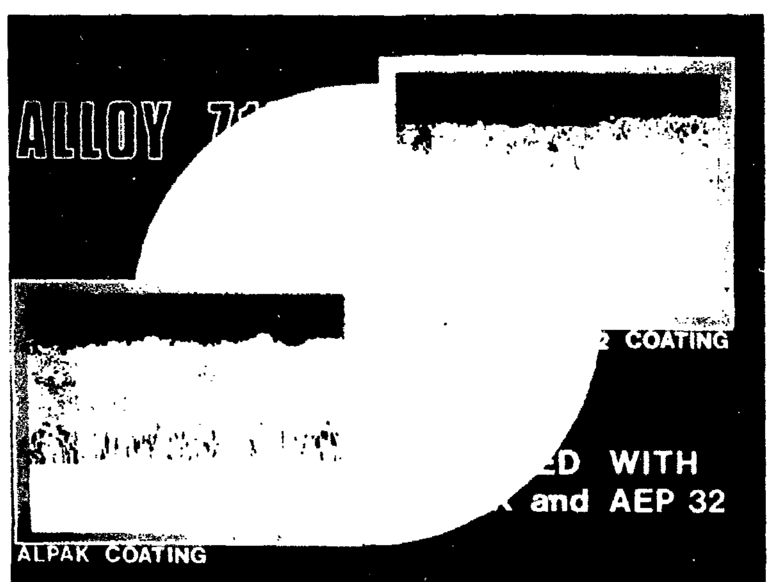

FIGURE 18. MICROSTRUCTURES OF ALPAK AND AEP-32 COATED 713C

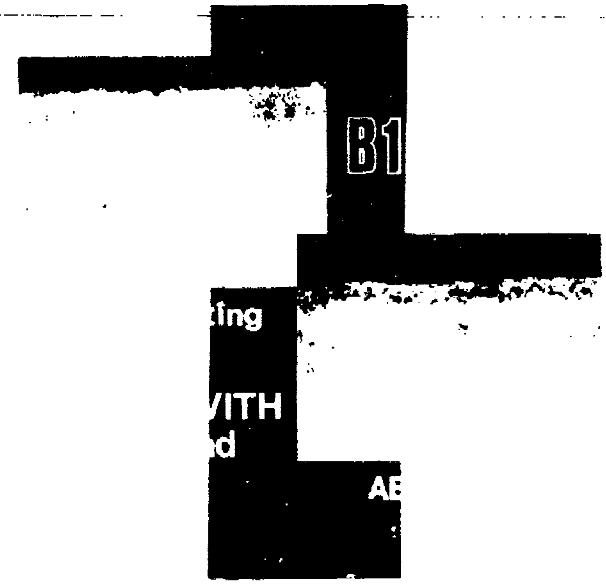

FIGURE 19. MICROSTRUCTURE OF PWA 73 AND AEP32 COATED B1900

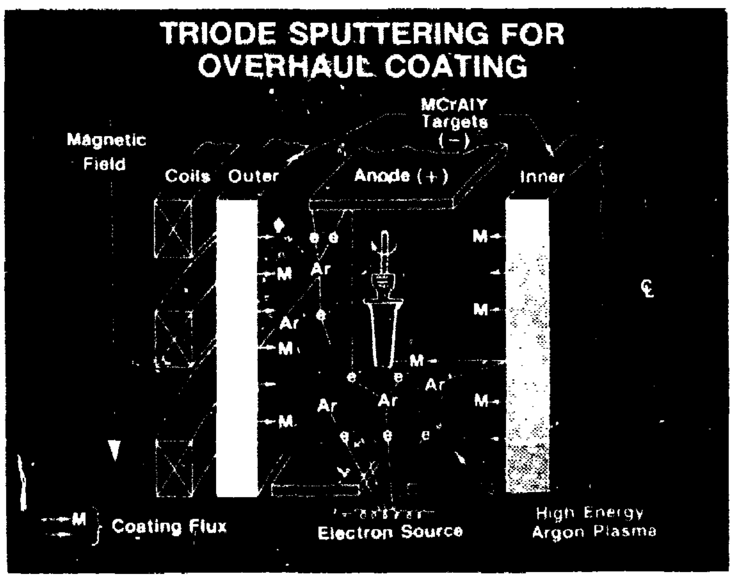

FIGURE 20. TRIODE SPUTTERING COATING PROCESS FOR DEPOSITING MCrAlY OVERLAY COATINGS TO TURBINE AIRFOILS

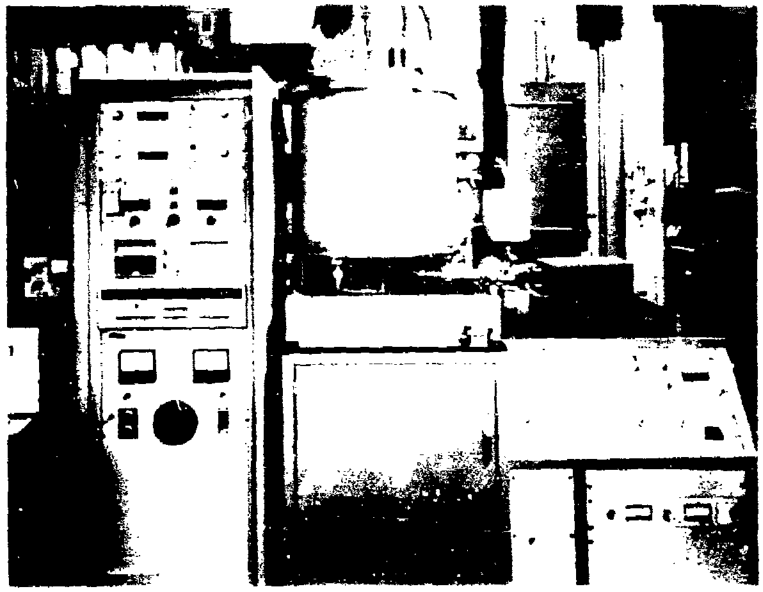

FIGURE 21. CATHODE SPUTTERING SYSTEM

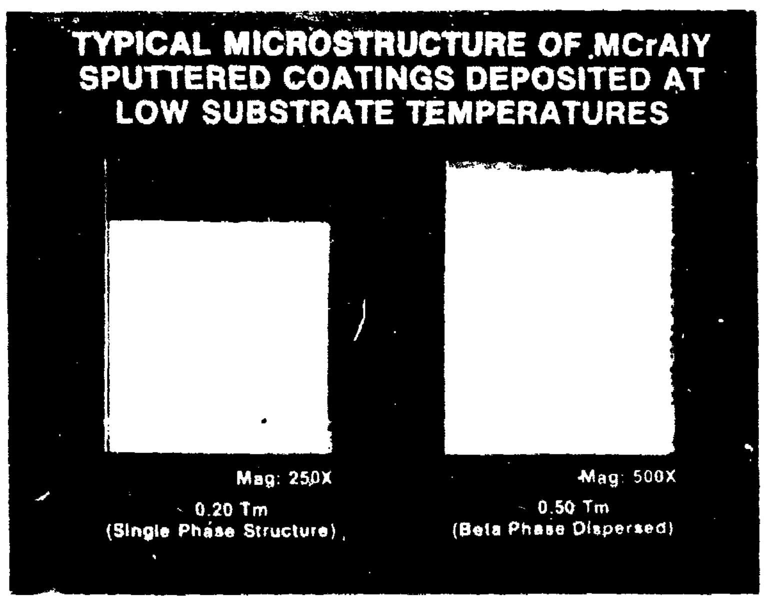

FIGURE 22. TYPICAL MOCROSTRUCTURE OF MCrAlY SPUTTER COATING DEPOSITED AT LOW SUBSTRATE TEMPERATURE

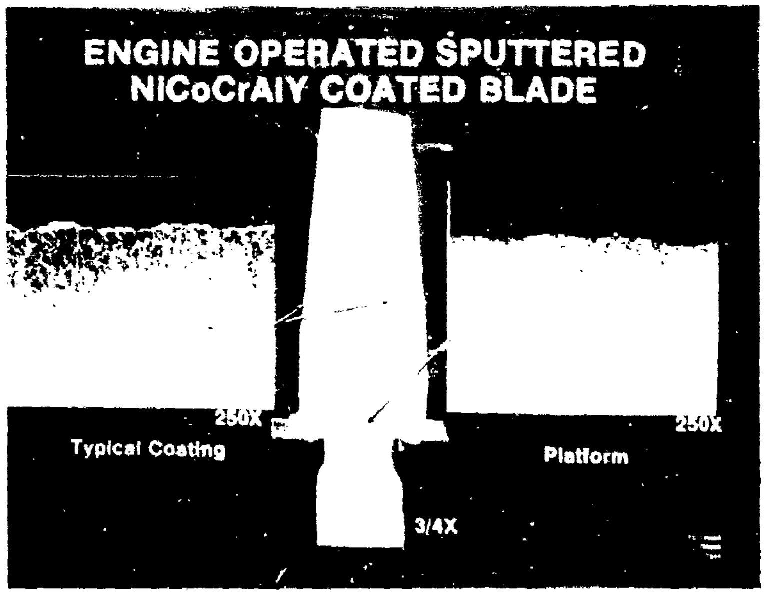

FIGURE 23. ENGINE OPERATED SPUTTERED NiCoCrAlY COATED BLADE

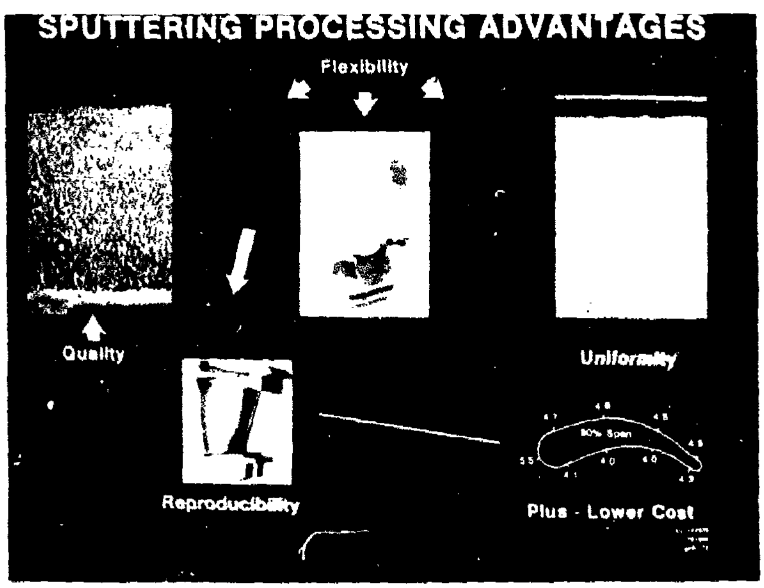

FIGURE 24. SOME ADVANTAGES OF THE SPUTTERING PROCESS

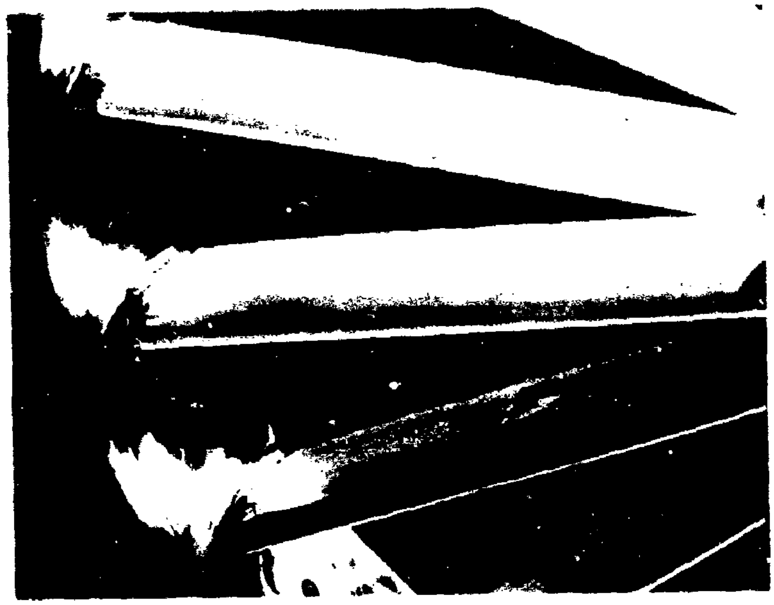

FIGURE 25. PROTOTYPE DAMPING WRAP ON TF30 INLET GUIDE VANE

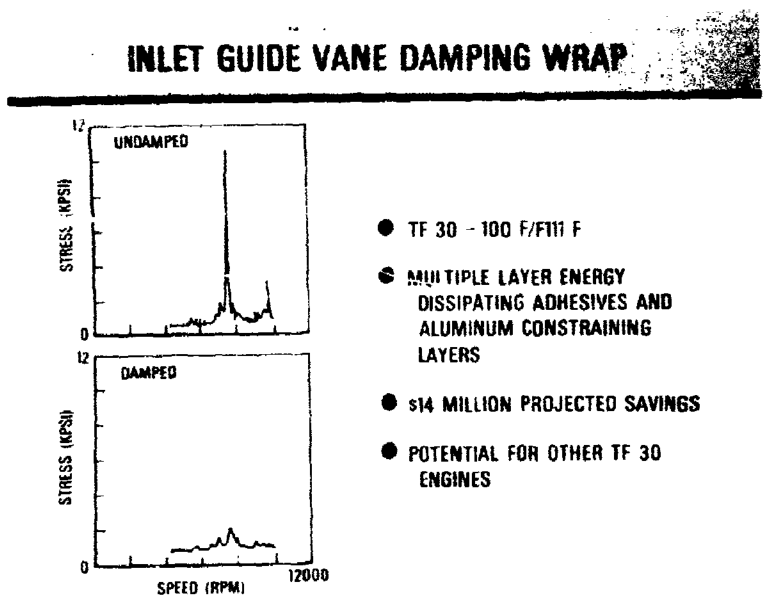

FIGURE 26. COMPARISON OF DAMPED AND UNDAMPED INLET GUIDE VANE

MAINTENANCE PROBLEMS IN GAS TURBINE COMPONENTS AT THE ROYAL NAVAL AIRCRAFT YARD FLEETLANDS

by

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SUMMARY

This paper deals, in a general sense, with the work of the engine repair facility at RNAY Fleetlands and describes the major problems found in the overhaul and repair of helicopter and marine gas turbines. Remedies for component reserviceability and developments to obtain longer service lives with a description of the techniques employed are discussed.

INTRODUCTION

The RN Aircraft Yard, Fleetlands, was established in 1940 for the overhaul of aircraft and engines for the Fleet Air Arm. From repairing of Swordfish and their internal combustion engines Fleetlands advanced to working on pure jet engines in 1954 and then increasingly with helicopter turbo-shaft engines. In 1967 Fleetlands became the tri-Service Helicopter Repair Centre and as the largest group within the Naval Aircraft Repair Organisation it is responsible for the repair and modification of helicopters, engines and components at various levels. Its other tasks include the repair of hovercraft and the Tyne and Olympus Marine Gas Turbines as Change Units (MGTCUs) for the Fleet.

These engines can be life expired or may require specific work to be carried out such as repair of foreign object damage, rejection for engine health monitoring reasons, or part life refurbish of the hot end of a marine gas turbine. In addition aero engines may be received for defect investigation or as part of service trials.

Military engines generally will have a lower time between overhaul than those in civilian usage. This reduced time can be due to the effects of rapid temperature changes in the hot end of the engine and the frequent change of power output during service. Military aircraft and helicopters in particular operate at lower altitudes and therefore are much more subject to the machinations of the environment than high flying civilian aircraft. These areas include operation in high humidity and salt laden atmospheres, making corrosion probably the greatest problem to overcome particularly within the Naval area. Sand provides in situ grit blasting and operation in tropical coastal areas produces all three, having dramatic effects on cold end components. At present, military aircraft have a low utilisation and this gives rise to other problems particularly that of corrosion before the engine would have otherwise completed its designated life. RN service engines are frequently compressor washed with fresh water for removal of salt which if done meticulously is very effective. If incorrectly carried out the effect can be worse since salt deposits would be washed into areas which operate at high temperature during service. Where deposits are sooty or oily, detergents are employed but there can be some adverse effects on some diffusion coatings.

The following paper gives a general description of engine overhaul at RNAY with a description of particular problem areas found in hot end components and their repair. Obviously many of the schemes employed have been developed by the engine manufacturers but an increasing proportion have been initiated and developed within our own organisation as is particularly the case with the older engines. No reference will be made to the compressor system but it should be noted that this area probably produces the greatest proportion of problems found and particularly in relation to saline corrosion. Examples of hot end component failures are given and although these cannot be described as maintenance problems they do follow from problems in operation and I feel it is appropriate to raise them at this conference.

GENERAL OVERHAUL PROCEDURE

The aim of the repair process is to restore a degraded engine whose life has expired to a state that allows it to complete a further life. This procedure basically consists of 6 major operations: strip, clean, inspection, rectification/salvage, build and test. Of these no reference will be made to build and test.

Since all parts of a gas turbine are vulnerable to damage or wear during operation each unit must be completely stripped. All the engine types are dismantled in a specific area by the same operators. The monitoring of components from each engine is controlled by a system of pallets; the components being segregated into specific groups in pallets thus allowing quick identification of parts and easy transportation over a system of roller conveyors. Once stripped and segregated the parts are dispatched to their own special treatment areas. The majority of parts will pass through a cleaning cycle, typically consisting of a paraffin wash followed by immersion in one of several degreasing agents, a fresh water wash before drying by compressed air or in a warm air cabinet. In addition, items may require the removal of oxidation products.

Over the years the Yard has been faced with significant problems in the removal of high temperature protective paints from compressor blades without damaging the blades. Progress has been made but it still remains a problem area. As great a problem is related to the cleaning of hot end blades due to the tenacious oxidation and combustion products. Hot caustic solutions are employed containing various additives to aid conditioning of the oxide and its removal. Hot processing temperatures are in the region of 90°C. The hot chemical processing has been found still to be inadequate and the use of ultrasonic cleaning associated with the hot chemical action produced beneficial results. More recent experiments using a vibratory polishing process allied with the hot chemicals appears to have produced even better results.

Once cleaned and dried the components are in a suitable condition to be examined for defects. The components are surveyed visually or with binocular assistance as necessary following fluorescent or dye penetrant crack detection processing. Engine blades are processed in batches in an automatic cleaning and crack detection plant using the fluorescent dye penetrant method.

Many of the initial stages of the turbine blades and nozzle guide vanes of our engines are pack aluminised. The cleaning of these components is markedly more difficult than those without the pack aluminising mainly due to the roughened surface of the aluminised component providing a better key. The subsequent fluorescent dye penetrant inspection of aluminium blades can give cause for some concern. It would always be preferable to remove the coating in order to check the integrity of the blade. This is carried out on Marine Gas Turbine blades and nozzle guide vanes but not on our aero engines. The small aero engine blades would lose too much material from stripping and re-applying any aluminised coating. Cracks have been detected in pack aluminised blades and there have been no spates of failures arising from this course of action.

Marine gas turbine blade and nozzle guide vanes due to the much poorer environment of saline atmosphere allied with lower quality fuel do suffer severely and suspect components are normally returned to the manufacturer for inspection and re-aluminising.

Following dye penetrant inspection the next stage is scrutiny of each engine component by a team of inspectors whose task is to reject those parts which fail to meet the standards laid down in the overhaul manual for that particular engine. Many rejected parts can be restored to a serviceable condition and are routed to various support facilities within the Yard such as the machine shop, plating shop, welding and thermal spraying groups etc.

The helicopter engines and marine gas turbine are dealt with separately due to the increased severity of environment in which the marine gas turbines operate.

HELICOPTER GAS TURBINES

(a) Combustion System

The combustion chambers of the older engines such as the Gazelle, Nimbus and Gnome are essentially manufactured from solid solution hardened alloys such as Nimonic 75 since they are readily produced in the form of sheet, possess good forming characteristics, and may be easily welded. However they do have only a limited high temperature strength and significant cracking is found in these older systems from stress raisers such as location slots, diffuser and cooling holes. Direct weld repairs or weld repairs by replacement are simple and involve argon arc welding with 80Ni-20Cr filler wire having previously thoroughly cleaned the cracked areas. Unclean joints will produce porous welds and can result in hot cracking. The arc gas is generally pure argon but Ar/H_2 mixtures have been used to advantage to reduce porosity levels and improve weld penetration for a given power setting.

In areas close to the actual flame thermal fatigue is a significant problem and there have been numerous failures in the older engines such as the early marks of the Gnome (Fig 1), (1). These have been somewhat improved by the replacement of various sections of this particular system with Hastalloy X giving improved high temperature properties than that of Nimonic 75.

Carbon erosion used to be a problem in the Gnome but has been essentially removed by modifications to the burner system, although we still see the occasional problem.

The modern combustion chambers are based on Nimonic C263 materials (Gem engine) a precipitation hardened alloy. Our experience to date has been that every combustion chamber exhibits distortion and severe cracking between cooling/diffuser holes (Fig 2). Weld repairs following the approved scheme results in weld centre-line cracking due to the residual stress present within the component. The added presence of oxidation and combustion products does not help the situation. The complete removal of these stresses by solution treatment prior to welding may solve this problem but may result in further distorting the component due to its complex shape and thin section. This process is being examined along with the use of a high temperature braze repair technique.

Fretting is a frequent occurrence at interconnectors and locating flanges. In the older engines these areas would typically be hard chromium plated and when fretted beyond acceptable limits repair would be by re-hard chromium plating when possible or more frequently by welding on a replacement part. Hard chromium plating is not the most suitable material for fretting resistance under these conditions since at temperatures above 450°C its hardness drops dramatically such that the thin coating is rapidly worn through resulting in localised wear occurring in the softer parent material, usually Nimonic 75. Re-plating then becomes impracticable. Plasma spraying using a nickel-chromium/chromium carbide composite is an effective repair to maintain design dimensions with a coating that maintains good hardness and wear resistance at high temperatures (2). The harder and more fretting resistant ceramic and tungsten carbide containing composites were not preferred because the operating temperature was such that a degradation of these coatings would occur.

Cracking and erosion of the holes on the Nimbus diffuser shown in Fig 3 is a frequent problem. A direct weld repair of the holes is not possible but replacement of the diffuser front can be readily

carried out. This can be performed by TIG processes but electron beam welding gives a faster result particularly when processed in batches. Greater reliability on weld quality is an added factor. It is anticipated that a similar technique can be employed in the Astazou IIIN diffuser which suffers from the same problem but after a longer life due to its greater material thickness (Fig 4).

(b) Turbine Section

(1) Turbine Blades

In general we experience relatively few problems with turbine blades where these blades are separate items. The gas generator blades are cast blades of SEL (Gnome and Nimbus) and MAR-M-246 (Gem) and are unlifed. All are pack aluminised and invariably exhibit good corrosion and erosion resistance. The blade tips however are frequently ground to final size during build and thus there is no protective coating at the blade tips. This lack of coating has not resulted in any subsequent damage other than in Nimbus engines and even then only in blades exhibiting overheating.

In the case of the Nimbus engine we have been experiencing high levels of blade rejection. A high percentage are rejected immediately for the burning of blade tips with a comparative number rejected for overheating (Fig 5). In this latter case sectioning of suspect blades is carried out to confirm the degree of overheating and to examine for any other defects, such as creep damage, blade cracking and any breaching of the aluminised layer. Where these defects do not occur the blades remaining of the stage may be subjected to a full heat treatment as a salvage. Grain boundary oxidation of the blade tips in the direction of the blade root is being found thus rendering the blades as scrap. The reasons for the overheating effects have not yet been established, but the modified Nimbus engine has generated problems in nozzle guide vane distortion and cracking, the distortion of the shrouds frequently producingtip rub, shown in Fig 6 (3).

(2) Turbine Discs and Bliscs

These components are lifed and generally will fulfil their designed lives before scrapping. Corrosion and overheating problems are very low in gas generator discs. However, certain components, notably the Gnome power turbine disc and shaft suffer very badly from corrosion pitting on the disc faces, fir tree roots and in the shaft bore (Fig 7). Trials in the use of the electrophoretic paint, PL199, on Gnome hovercraft engines produced excellent results in that no further corrosion pitting or defects developing from existing pits were found. However, this potential solution having been found, material modifications from REX448 to A286 have been instigated and this is expected to resolve the problem.

The bliscs (turbine blade and disc manufactured from the solid) are only found in the Nimbus (pre-Mod 719) and Astazou IIIN engines. We still see occasional failures in the 2nd stage turbine of the Nimbus (pre-Mod 719) due to fatigue failure initially in the trailing edge of the blade. All of the Nimbus blades are pack aluminised and give satisfactory performance from a high temperature corrosion/oxidation viewpoint.

As yet little can be said on our experience with the Astazou IIIN as we have only just started the overhaul of this engine system.

(3) **Nozzle Guide Vanes**

The nozzle guide vanes are manufactured as a casting or more generally as a fabrication with vanes welded or brazed into the shrouds depending upon its operating temperature. The 1st stage nozzle guide vane is manufactured from the higher temperature super alloys and would be pack aluminised, the following cooler stages generally of Nimonic 75 fabrication with or without aluminised coatings.

On every engine that we overhaul cracking is always present, usually in the blade/shroud joints and blade trailing edges. Dependent upon the position, size and morphology of the cracks they can be tolerated, examples of which are shown in Figs 8 and 9. The cracks are chiefly due to thermal effects such as thermal fatigue cracks on trailing edges or insufficient strength at brazed joints, or due to engine and aircraft vibration. Erosion and impact damage does occur but in general this is not too serious.

Repair work may be conveniently split into those components which are aluminised high temperature super alloys and those of the cooler non-aluminised Nimonic 75 components. Until recently RNAY has been restricted to just carrying out argon arc weld of cracks in the parent material of N75 components. Any braze repair work would have been returned to the manufacturer. Any such vacuum brazing repair was restricted solely to cracks in the braze. Where cracks branched into the parent material these components would be scrapped.

Fleetlands has recently been carrying out its own programme on the use of high temperature braze repair using Ni-based braze materials. Pre-cleaning of the component is most important in order to remove combustion and oxidation products within the crack. This can be accomplished using the standard cleaning methods such as immersion in hot alkaline solutions, acid pickling or grit blasting followed by a high temperature anneal in dry hydrogen or high vacuum. This is then followed by vacuum brazing using Ni-based alloys and a longer cycle time than normally employed in standard brazing practice. The aim being to carry out a diffusion brazing procedure in order to reduce and hopefully remove centre line phases which would be detrimental to the joint mechanical properties. The joint clearances of cracks vary from wide at the mouth to zero at the tip and usually there are multiple cracks at varying orientations. This obviously creates problems initially on cleaning the crack root but also on selection of the

braze materials employed. The free flowing brazes required to penetrate to the root of the crack may not be suitable for the crack mouth. Various combinations of braze mixtures and filler materials have been employed to overcome this and also to be compatible with the braze material used in manufacture. A number of nozzle guide vanes have been repaired using the above technique to repair cracks and foreign object damage shown in Fig [illegible] and it will be interesting to observe their condition through operational service.

With regard to the higher order super-alloys, eg Nimonic 90 and C263, which contain increasing amounts of Ti and Al, cleaning of the components becomes increasingly difficult as the levels of Ti and Al increase since these elements are essentially employed to increase high temperature oxidation resistance as well as improving the high temperature mechanical strength through precipitation of sub-microscopic γ' [cuboids of Ni_3 Al, Ti] . Certainly a vacuum cleaning operation will not be sufficient to clean the cracks and even extremely dry hydrogen may be insufficient. The use of a more reactive reducing species may well be the answer. Our work on this is continuing.

(c) Exhaust Casings

The gas temperature in this area of a turbo-shaft engine tends to be relatively low and materials such as 18/8 austenitic stainless steel or Nimonic 75 are employed. The casings are of complex shape to deflect gases away from the aircraft fuselage and rotor which when allied to the thin gauge materials used inevitably produce distortion and cracking. We have developed over the years various welding and patching schemes other than those of the manufacturers for weld repair and replacement of defective parts. Frettage is also a significant problem on flanges and we are again developing our own schemes to combat this either by replacement or by the use of plasma spraying for building up and providing frettage resistant coatings.

MARINE GAS TURBINES - TYNE AND OLYMPUS

a) Combustion Section

A major source of problems from the earliest days of the marine gas turbines have been related to the fuel and combustion system (4). In both engines but particularly the Olympus the RN service requirements for smoke levels were never achieved using the standard multiflare can (Olympus) with the original burner. Extensive modifications resulted in greatly improved smoke levels but at the cost of increased complexity. This in turn resulted in problems for GTCU interchangeability when matching spare GTCUs with the progress of module specifications. Combustion can life has been greatly improved from 1000 hours for the original to 2500 hours with the developed phase 1 can, which when combined with the improved 'EX' burner design gives generally satisfactory performance.

Typically problems related to the can are of cracking in the swirler vanes which may be repaired by high temperature vacuum brazing, as are the occasional cracks at diffuser holes. The inlet and outlet flanges are coated with a detonation sprayed tungsten carbide/cobalt composite powder and give satisfactory performance, although a plasma spray coating using the same composite powder may be a satisfactory alternative at lower cost and quicker turn round times. Until this summer RNAY had not received the uprated version of the Tyne RM1A, the RM1C.

Typical problems of the early RM1A engines would include distortion of the Nimonic 75 combustion chamber, fretting at hard chromium plated interconnectors and the serious build up of carbon within the chamber which, when released, resulted in severe carbon erosion of downstream components. Improvements in burner design has greatly reduced this problem and given a resultant increase in engine life.

The uprated RM1C produces higher temperatures resulting in greater distortion and cracking of the combustion chambers (4). Material modifications to the stronger C263 alloy failed to cure this problem. Subsequent developments by the manufacturer using Nimonic 86 with thermal barrier coatings are proving most successful to such an extent that engine lives of 5000 hours are now expected.

Olympus burners suffer from severe corrosion in the barrel area. The component is a 0.4%C item forging with a 10-15 μm nickel plated coating for corrosion protection. When exposed to sulphurous atmosphere (the sulphur mainly originating from the Dieso fuel) free sulphuric acid is produced. This, together with the fine salt spray found in marine engines produces localised pitting corrosion, followed by corrosion beneath the nickel layer then subsequent severe attack. A weld repair scheme developed at Fleetlands salvages these components and the use of ceramic barrier coating may well significantly reduce this problem.

(b) Turbine Section

As stated above, carbon erosion was a major problem on the early Tyne RM1A engines as shown in Figures 11 and 12. Carbon erosion still occurs but is not now considered to be a major problem.

With regard to hot end corrosion the literature is full of information of turbine blade deterioration under severe chloride ingestion and in the presence of sulphur, usually from the fuel. Since this is a major field of its own it is not proposed to discuss the mechanisms here. Up to the present time only pack aluminised coatings have been employed in RN service other than in service trials. Recent changes to the use of platinum aluminide coatings are expected to give longer

service lives with a less brittle coating. Corrosion effects are markedly temperature dependent - hot end corrosion is no different. There certainly appears to be 2 forms of corrosion taking place (shown in Figs 13 and 14), each in its own temperature regime, and no coating yet developed has provided adequate protection against both forms. The implications from reported research work carried out in this field are the use of high chromium and intermediate aluminium levels in coating and high Ti/Al ratios in the parent material.

The range of coatings from the standard pack aluminising, through precious metal variants of aluminising and the more recent M Cr ALY coatings and their applications by diffusion or overlay techniques is very wide and there is no doubt that this is an important area for long term development. This is especially so as longer service lives of gas turbine systems are demanded.

NON-DESTRUCTIVE TESTING

The standard techniques are employed in the examination of hot components with generally visual, dimensional and dye penetrant examinations with judgements based on accumulated experience backed up by resident engine manufacturers representatives. As stated above, however, the one group of components which does give some cause for concern are pack aluminised blades due to the difficulty in cleaning the components without damaging the aluminised coating. Cracks have been detected in such blades and as there has been no spate of failures we are reasonably satisfied although not complacent.

One further technique that we have recently started specifically on turbines of the Astazou IIIN involves replicating specific areas of the component. These areas are mechanically and electrolatically polished to remove oxidation and combustion products and leave a highly polished surface. A plastic replica is taken of the surfaces and is then examined between glass slides under a microscope for defects which may not be found by dye penetrant techniques. These would include fine impact damage, corrosion, surging, creep and fatigue cracking. Mechanical damage on the curvic coupling will also cause grain boundary defects, but this is usually coarse enough to be detected by dye penetrant techniques. A certain amount of rework is allowable and this would be followed by a repeat replication of the area to ensure complete removal of the defects. The process is time consuming and requires operator experience but can show up significantly more detail than otherwise would have been obtained and allows expensive components to be re-used whereas they may have been scrapped.

Where components have been salvaged by welding, brazing or thermal spraying techniques there is obviously a need to know that the work has been satisfactorily carried out. Such components would be subjected to dye penetrant examination and also X-radiography when required and where it is practicable. All welders are required to produce frequent test pieces which are subjected to critical examination, and if not satisfactory can result in the welder losing his "licence" to continue.

The testing of thermally sprayed coatings is much more difficult since the only real test is one involving destruction of the component. The use of test pieces is again required and it is essential that these be carried out under the same spraying conditions and those of component for them to be meaningful. This is particularly important where abradable seals, such as nickel/graphite, and boron nitride/cement powders, are employed. Too soft a coating will result in excessive coating wear; too hard a coating will result in blade tip wear or chipping of the coating resulting in impact damage downstream. Metal carbide containing coatings (eg WC, Cr_2C_3) again would be tested to ensure proper fusion, distribution and morphology of the carbides, porosity levels and obviously for a clean coating/substrate interface by metallographic examination.

Bond coats of Ni-Al are normally subjected only to an occasional check to ensure quality of the coating since these materials offer a wide tolerance in operating conditions during spraying.

Wherever possible the process should be automated to give consistency of spraying conditions between components.

Until recently there was no non-destructive test of the coating other than visually. Franke and de Gee (5) and subsequently Reiter (6) have developed ultrasonic methods to establish areas of poor contact between the coating and its substrate and within the coating itself and although this cannot be related to adhesive strength must give confidence in the technique.

CONCLUSION

In conclusion to what is a somewhat general paper RNAY Fleetlands experiences specific problems in each of the engine systems it is required to overhaul and these tend to be highlighted in component shortages of particular engines, several of which are over 20 years old and still have to be kept in service. If specific areas must be highlighted these would include:

(a) Cleaning of turbine blades and especially coated blades prior to inspection.

(b) The problems associated with the repair of aero engine nozzle guide vanes and to a lesser extent, combustion chambers, which are individually expensive and almost always in short supply. Research carried out has produced successful results in the repair of the Nimonic 75 based NGVs but as the higher order Nimonic alloy assemblies certainly require more work.

Like any establishment working with highly complex pieces of machinery we can be subjected to the whims of supply and it is hoped that this paper gives some indication of our problem areas in the field of this conference and developments that we have adopted or developed ourselves in order to overcome those difficulties.

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(6) Reiter, H, University of Bath, Private Communication.

ACKNOWLEDGEMENTS

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This paper expresses the opinions of the Author and does not necessarily represent the official view of the Ministry of Defence.

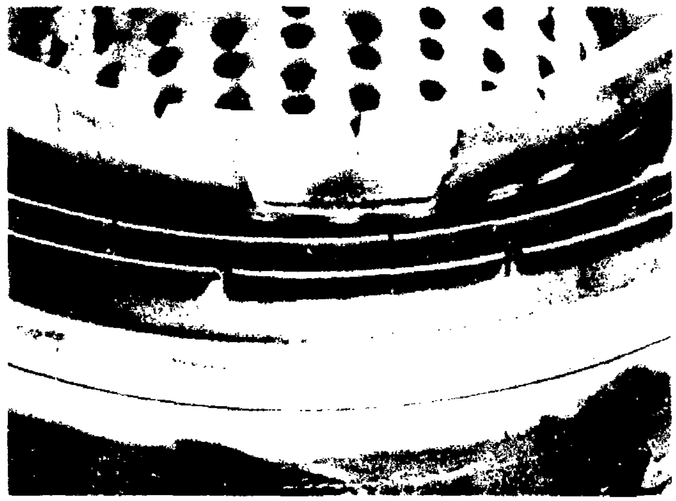


Figure 1. Failure in a Gnome combustion chamber at the Stage 1 nozzle guide vane locating shroud.

Figure 2. Cracking on the inner diameter of the Gem combustion chamber and burning of air shrouds.

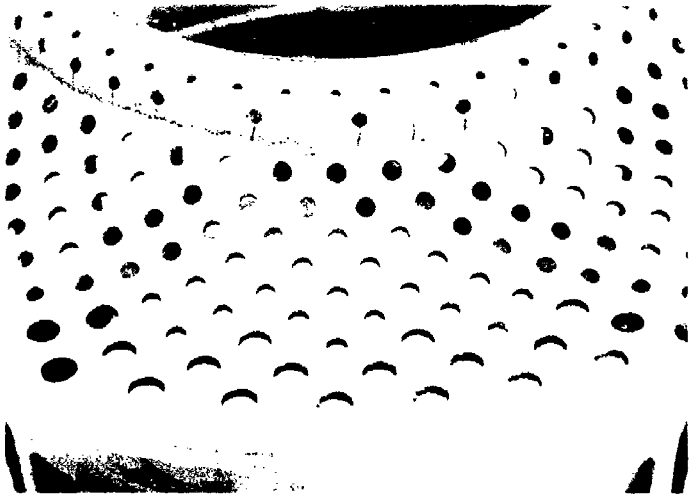


Figure 3. Thermal fatigue cracking from holes in the [illegible] diffuser assembly.

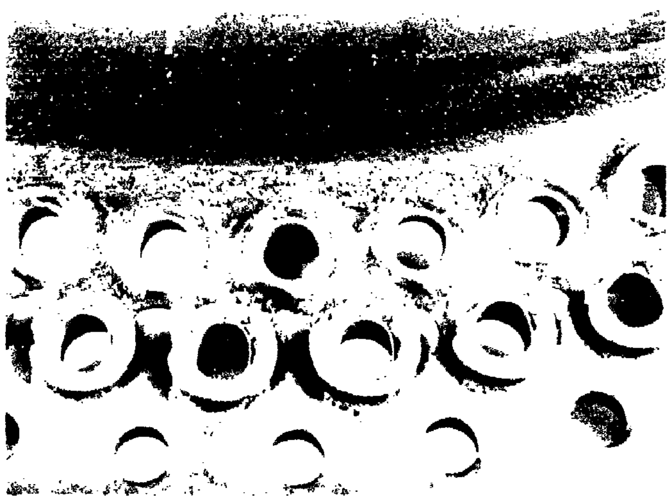


Figure 4. Erosion and cracking on the Artazon diffuser assembly.

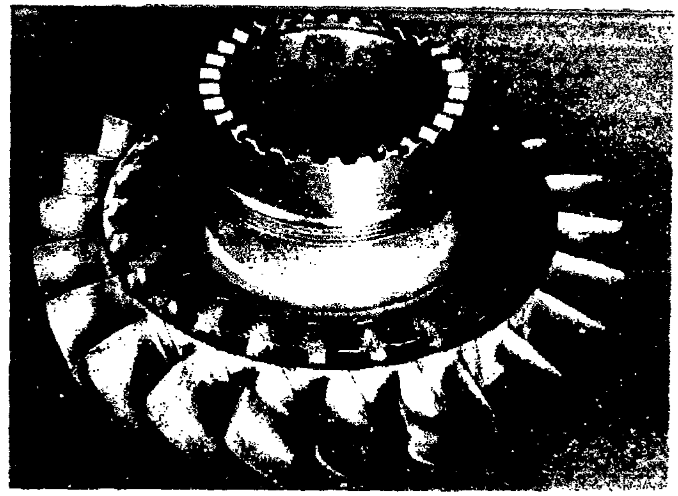


Figure 5. Overheating of Stage 1 turbine blades of the Nimbus (post mod [illegible]) engine shown by the 'blued' appearance in the upper half of the aerofoil.

Figure 6. Turbine tip rub on the Nimbus Stage 2 nozzle guide vane and staining of the outer shroud.

Figure 7. Corrosion pitting of the disc face and fir tree roots of the [illegible] power turbine [illegible] and shaft.

Figure 8. Thermal fatigue cracking and loss of pack aluminized coating [illegible] nozzle guide vane.

Figure 9. Cracking adjacent to the outer shroud braze in the Nimbus Stage 2 nozzle guide vane.

Figure 10. High temperature braze repair of cracks in the braze, guide vanes and foreign object damage of the component shown in Figure 9.

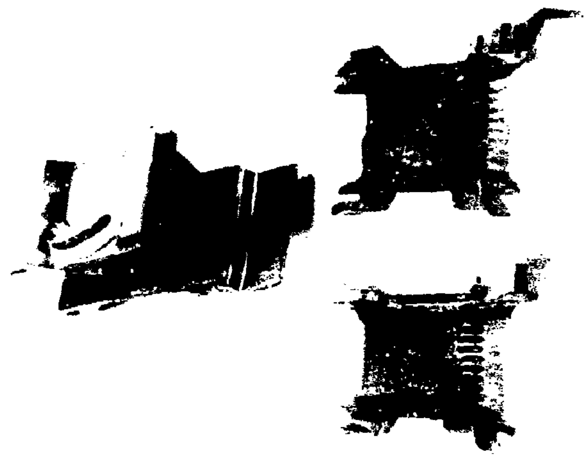


Figure 11. Carbon erosion of Marine Tyne engine nozzle guide vanes at locating shrouds (left), aerofoil face and bending edge (top right) as compared to a blade in good condition (bottom, right).

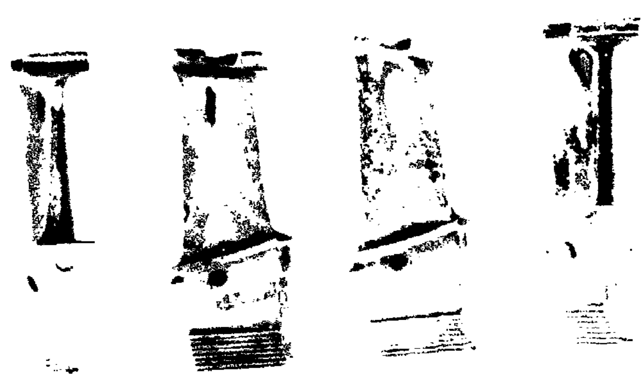
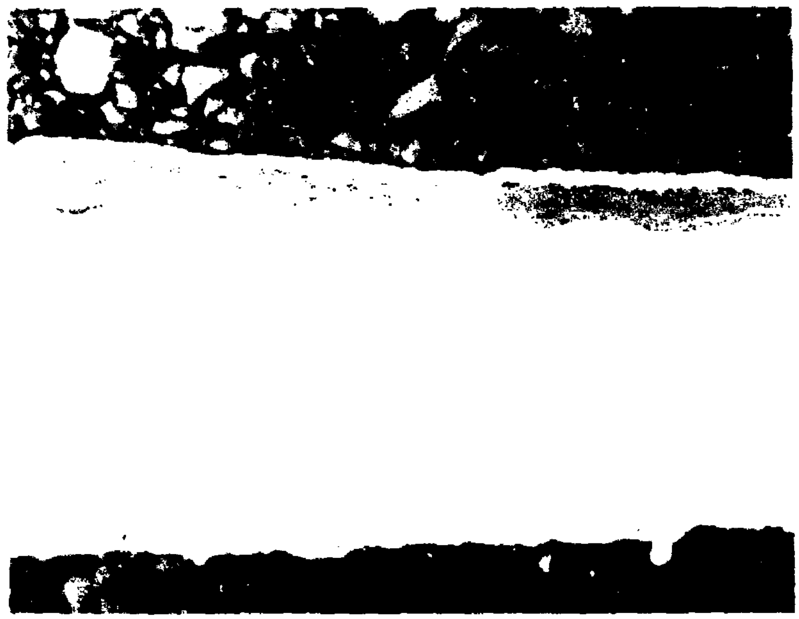


Figure 12. Carbon erosion of Marine Tyne HP turbine blades at 2300 hours life (right) and 3000 hours life (left).



Low Temperature Sulphidation

High Temperature Sulphidation

Figure 13. Sections of corroded Tyne HP turbine blades showing the 2 forms of sulphidation corrosion, the layered appearance of the low temperature form and the severe grain boundary attack of the higher temperature form.

Magnification x 130

MAINTENANCE EXPERIENCE WITH CIVIL AERO ENGINES

by
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SUMMARY

In civil airline operation the maintenance and fuel costs are a major item of concern and subject to close control in order to survive the hard competition between the airlines. Preventive maintenance of gasturbine engine High Temperature parts is mostly limited to visual examination and/or performance EGT trend shift triggers, aiming at avoiding extensive secondary damage. The trends in maintenance concept developments are reviewed, indicating the constant activities to optimize the maintenance cost of the propulsion system. As a result of the escalating trends in material and fuel prices, the presently applied maintenance concepts require a more sophisticated condition control in order to comply with the need to find the optimum operating time of each individual engine. With the introduction of the new generation of civil aircraft (Airbus A310) a mutual goal between the engine manufacturer and the airline has been defined to develop mathematical programs, based on actual recorded conditions, in order to control the behaviour of the engine, aiming at an optimum use of the propulsion system.

INTRODUCTION

For a civil operator like KLM it is of a prime objective to sell passenger and freight transportation, in a fast and convenient way, competitive with other means of transportation but also economically justified and aiming at a financial positive end result. From this point of view, maintenance and fuel to be used is considered to be a negative contribution to the ratio of cost to revenue. The aircraft propulsion systems are, because of deterioration and fuel burn reasons, a major contributor to the total utilization cost of the airplane and thus subject to continuous improvements in order to obtain the highest propulsion thrust for the lowest fuel burn and weight penalty.

Using gasturbine engines for aircraft propulsion a balanced compromise has to be derived between two extremes:

- Based on the Brayton cycle a high turbine inlet gas temperature is a prime contributor to overall engine efficiency, however, this has a direct impact on material cost and durability of the engine.
- Operating with a low turbine inlet gas temperature allows less advanced materials to be used, however, the resulting efficiency loss has also a large impact on operating costs.

The practical side of the problem is that the past design efforts to build more powerful engines are superseded by the design aim to have the most efficient engine against the lowest utilization and maintenance cost.

MATERIAL AND FUEL BURN COSTS

Both material and fuel costs have escalated substantially in the past seven years, resulting in proportionally higher operating costs.

Materials:	Cobalt	+700%
	Tantalum	+800%
	Titanium	+400%
	Chromium	+300%
Fuel	: Jet A1	+700%

The composite price index in the same time period increased by only + 100%.

Future trend developments are hard to forecast, e.g. a fuel price increase between 12% and 20% a year may be expected. Nevertheless the cause of the price increase is shortage of the major materials on the international market, and this condition only tends to get worse. The shortage on the market led to considerations as to which strategy has to be followed. Options are:
- scrap parts reclamation for recycling.
- selected use of materials
- development of alternatives.
- expansion of exploration.

There is not yet an active program developed for recycling of hot section parts, because it is not yet clear who should initiate this program. The development of alternatives might be the best long term solution.

The direct and related cost increase is passed onto the final users of the materials, in this case the operator, who in turn is not able to charge it completely to the passengers because of restrictive IATA rules and a strong competition between the airlines. The answer to this phenomenon is to lower the overall cost per passenger by utilizing the most advanced aircraft/engine types, tailored for a special mission. The present effort is concentrated on the fuel burn aspect, because this has a direct effect on operating cost. The raw material cost however, is of secondary importance since it contains approx. 3% of the final manufactured part price. In an attempt to lower the total SFC, the potential metal temperature and/or creep strength of the high press turbine improved by utilizing the latest metallurgical developments such as directionally solidified and even single crystal turbine blades, blades with a wire-like Tungsten reinforced framework, powder metallurgy and composites.

The use of these advanced techniques however, tends to form a vicious circle since the same basic materials are used, even in a more complicated composition and often more difficult or not at all repairable; resulting in higher engine part costs. The latest development, the use of ceramic materials is not yet ready for commercial introduction. However, the operator is interested in the capabilities of those materials since another improvement in SFC might be expected because of film cooling deletion and higher turbine inlet temperatures allowed, together with a relative cheap basic material. Also for the combustor application (major cobalt consumer) ceramics might be a future development.

MAINTENANCE CONCEPT DEVELOPMENTS

With the commercial introduction of the gasturbine engine for civil aircraft propulsion, the principles of the military maintenance concept were adopted. This hard time concept started out with the requirement to overhaul the complete engine every [illegible] operating hours. Complete engine overhaul was applied up to 1965, and replaced by a [illegible] operating hours for the less severe degrading cold section than for the more critical hot section.

Sectional engine overhaul was applied in the 1965-1969 time period and replaced by an engine module-oriented shopvisit concept. This concept, still based on hard times, was dedicated to the sequence of the shopvisit, e.g.: the first shopvisit inspection requires inspection and repair of the turbine section and visual inspection of the remaining engine sections. The second shopvisit calls for visual inspection of the low pressure compressor and overhaul of the remaining sections. The third shopvisit requires overhaul of the low pressure compressor, inspection and repair of the turbine section and a visual inspection of the remaining engine sections This sequence was repeated up till the sixth shopvisit, when a complete engine overhaul was applied and the sequence reinitiated.

With the introduction of wide-body aircraft in our fleet in 1972, the maintenance approach was changed to an "on-condition" modular engine maintenance concept, based on engine-hour thresholds for each individual engine module at the moment of shopvisit. On-condition maintenance, which is still in use, is based on an active line maintenance monitoring and inspection program aiming at detecting discrepancies in an early stage, thus avoiding major secondary damage. The following criteria for engine removal apply:

- Accute (Most flight crew triggers confirmed by a maintenance check)
- Deferred (Most maintenance inspection triggers with no direct failure risk)
- Other (Convenience reasons, campaigns or FAA limits if any)

Upon removal the workscope is based on individual module thresholds each with an upper and a lower threshold limit with a fixed delta. The escalation of the threshold is based on a number of two samples inspected and overhauled, selected from the engine inventory exposed for shopvisit between the upper and lower threshold and/or two samples with a forced inspection and overhaul requirement whenever the upper threshold is passed. All shopvisits without a sample requirement will be treated according to workscope compiled by an Engine Steering Group considering shop incoming inspection results, line maintenance reports, operational trend data and modification requirements. Above mentioned procedure is developed from past experience gained by the operator and an active design approach of the engine manufacturer in constructing a modular engine suitable for this maintenance concept.

This present maintenance concept does not provide the optimum utilization of the individual engine for it was assumed that the longest on-wing time (low shopvisitrate) also resulted in the lowest overall maintenance cost. This, however, was in general true for the time period in which no material and fuel shortage was evident. Two conditions require close control in order to determine the optimum workscope for the engine.

1. Relation SFC versus shopvisit cost at various shopvisitrates.
   In general the repair cost tends to go down with a low shopvisitrate, however, the fuel cost tends to increase under the same conditions.
2. The potential time to go for each engine is a function of the life limiting parts margin build into the engine.
   A small margin results in relatively low shopvisit parts costs, for only the limiting parts have to be removed. On the contrary a high shopvisitrate and the returning basic handling costs results. However, a large margin allows a lower shopvisitrate, but the material cost will rise sharply as a result of increased material consumption (less utilization of part life.)

The optimum of the above mentioned conditions is not a standard or fixed figure, but a function of the engine performance and material condition of each individual engine.

Future engine maintenance requires a detailed health monitoring and mission analysis in order to create a computer model capable of controlling the optimum life of each engine (module), taking into account the actual fuel prices and material costs. Above mentioned optimalization procedures may cause a change in the on-condition maintenance concept, by acting on predicted conditions rather than actual failures.

LIMITATIONS IN INSPECTION CRITERIA

The condition monitoring of the engine is based on two routines, both aiming at an optimum use of the propulsion system and life limiting parts.

- Performance control:
  - Observation
  - Acceptance testing
  - Trending against a baseline
- Mechanical integrity:
  - Visual checks
  - Borescope inspections
  - Oil analysis
  - Non-destructive testing
  - Radio-isotope inspections
  - Ultrasonic inspections

Condition monitoring of the high temperature parts is limited to detecting exhaust temperature trend shifts at a reference power setting and visual observation, mainly by using borescope equipment as an aid. Considering the trending as an early warning feature in a more detailed visual inspection, it may be concluded that the mechanical integrity of the critical high temperature parts are only as severely inspected as the visual capability of the inspector allows. This is no problem when inspecting for obvious defects such as missing/broken or burned/heavy eroded engine parts, however, small cracks and minor defects in critical locations are hard to detect. Even for the shopvisit routines restrictions are encountered with respect to determining the remaining thickness of the thermal coating on the turbine blades and vanes. The present procedure to strip and recoat the turbine blade and vane surfaces at a certain threshold value gives no information with respect to the real coating condition as it is. However this recoating process may have an impact on the remaining fatigue and rupture strength.

From an operational point of view it might be concluded that the less severely an engine is used, the lower the fatique chance and deterioration will be. Because the take-off mode of the aircraft has the largest impact on the life of the engine, derating was introduced as a first attempt to comply with the objective to conserve on maintenance cost. Derating is presently applied whenever feasible. However, it is not yet an active tool to extend the useful life of the life limited components of each individual engine.

FUTURE CONTROL OF HIGH TEMPERATURE PARTS

Since appropriate methods are not yet available to control the basic engine behaviour in a more sophisticated way, a mathematical method to count the severity of the cycles in a flight mission was developed for military use. Although less severe in respect to the military missions, the routes flown by the civil operator could be subjected to the application of those mathematical routines. The first attempt to utilize this approach on civil aircraft, conducted by SAS and supported by Pratt & Whitney, is based on a mission analysis program using statistical data. The severity factor defition, based on actual data has not yet been evaluated, but future on-board recording systems like the AIDS on the AIRBUS A310 are defined in such a way that all corner point conditions are recorded and thus available for more detailed severity analysis.

For the A310 the engine manufacturers made commitments to develop an integrated refined cycle counting program as part of the engine condition monitoring package. The operator needs the manufacturer's know-how concerning the significant temperature and loading characteristics of the selected materials and the resulting impact from exposure to operating conditions in terms of low cycle fatigue and thermal cycles. Control of the mechanical condition of the engine parts, combined with a module oriented performance condition monitoring program such as Gas Path Analysis applied "on-wing", wich is also a mathematical program providing valuable information in respect to the actual engine, module and sensor condition, provides an overall assessment of the engine and its parts condition.

In the operators opinion these new mathematical approaches are a potential feature to control the advanced-design and extremely costly high temperature parts and fuel burn in order to utilize the optimum life of the engine.

It may be concluded that the civil operators are forced to become aware of the need to create or obtain an effective maintenance tool to further control the behaviour of the engine in order to optimize the material uses and fuel consumption, all aiming at the objective to keep air transportation economically justified.

ENGINE COMPONENT RETIREMENT FOR CAUSE

by

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ABSTRACT

Traditionally, cyclic life limited gas turbine engine components have been retired from service when they reach an analytically determined lower bound life where the first fatigue crack per 1000 parts could be expected. By definition, 99.9% of these components are being retired prematurely as they have (as a population) considerable useful life remaining. Retirement for Cause (RFC) is a procedure which would allow safe utilization of the full life capacity of each individual component. Since gas turbine rotor components are prime candidates and are among the most costly of engine components, adoption of a RFC maintenance philosophy could result in substantial engine systems life cycle cost savings. Two major technical disciplines must be developed and integrated to realize these cost savings: Fracture Mechanics and Nondestructive Evaluation. This publication discusses the methodology and development activity required to integrate these disciplines that provide a viable RFC system for use on military gas turbine engines. The potential economic benefits of its application to a current engine system are also illustrated.

INTRODUCTION

Historically, methods used for predicting the life of gas turbine engine rotor components have resulted in a conservative estimation of useful life. Most rotor components are limited by low cycle fatigue (LCF), generally expressed in terms of mission equivalency cycles or equivalent engine operation hours. When some predetermined life limit is reached, components are retired from service.

Total fatigue life of a component consists of a crack initiation phase and a crack propagation phase. Engine rotor component initiation life limits are analytically determined using lower bound LCF characteristics. This is established by a statistical analysis of data indicating the cyclic life at which 1 in 1000 components, such as disks, will have a fatigue induced crack of approximately 0.8 mm length. By definition then, 99.9% of the disks are being retired prematurely. It has been documented that many of the 999 remaining retired disks have considerable useful residual life. Retirement for Cause would allow each component to be used to the full extent of its safe total fatigue life, retirement occurring when a quantifiable defect necessitates removal of the component from service. The defect size at which the component is no longer considered safe is determined through nondestructive evaluation and fracture mechanics analyses of the disk material and the disk fracture critical locations, the service cycle, and the overhaul/inspection period. Realization and implementation of a Retirement for Cause Maintenance Methodology will result in system cost savings of two types: direct cost savings resulting from utilization of parts which would be retired and consequently require replacement by new parts; and indirect cost savings resulting from reduction in use of strategic materials, reduction in energy requirements to process new parts, and mitigation of future inflationary pressure on cost of new parts.

RETIREMENT FOR CAUSE METHODOLOGY

Philosophy — The fatigue process for a typical rotor component such as a disk can be visualized as illustrated in Figure 1. Total fatigue life consists of a crack initiation phase followed by growth and linkup of microcracks. The resulting macrocrack(s) would then propagate subcritically until the combination of service load (stress) and crack size exceeded the material fracture toughness. Catastrophic failure would result if the component had not been retired from service. To preclude such cataclysmic disk (and possibly engine) failures, disks are typically retired at the time where 1 in 1000 could be expected to have actually initiated a short (0.8 mm) fatigue crack. By definition when 99.9% of the retired disks still have useful life remaining at the time they are removed from service. Under the retirement for cause (RFC) philosophy, each of these disks could be inspected and returned to sevice. The return-to-service (RTS) interval is determined by a fracture mechanics calculation of remaining *propagation* life from a crack just small enough to have been missed during inspection. This procedure could be repeated until the disk has incurred measurable damage, at which time it is retired for that reason (cause). Retirement for Cause is a methodology under which an engine component would be retired from service when it had incurred quantifiable damage, rather than because an analytically determined minimum design life had been reached. Its purpose is *not* to extend the life of a rotor component, but to utilize safely the full life capacity inherent in that component.

This philosophy in itself is not new, and has been used successfully in the past for low-stressed components which were generally durability, not fracture, critical. For this discussion, fracture-critical components are those components whose failure would likely result in engine power loss preventing sustained flight through single or progressive parts failure. Durability critical components are those parts whose failure would result in a significant maintenance burden but would not likely result in a flight safety problem. The advent of high thrust-to-weight engines with sophisticated (and costly) rotor components made from critical strategic materials, coupled with advances in the understanding of the fracture processes, and improvements in nondestructive inspection, has prompted a renewed interest in applying this philosophy to fracture critical components.

Residual Life — For simplicity, this section will consider component life limits which have been determined from the crack initiation characteristics of the specific disk material, and will not address the problem of intrinsic crack-life defects. While damage tolerant concepts are utilized in some instances to establish life limits, the majority of components in current gas turbine engines have had life limits set by an initiation criterion.

All fatigue data have inherent scatter. The data base used for design life analysis purposes must be applicable to all disks of a given material, and therefore includes test results from many heats and sources. Data are treated statistically as shown schematically in Figure 2. The distribution of life, defined as the number of cycles necessary to produce a crack approximately 0.08 mm long, is obtained for a given set of loading conditions (stress/strain, time, temperature). As can be seen, the $\pm 2\sigma$ bounds, which contain 95% of the data, may span two orders of magnitude in fatigue initiation life.

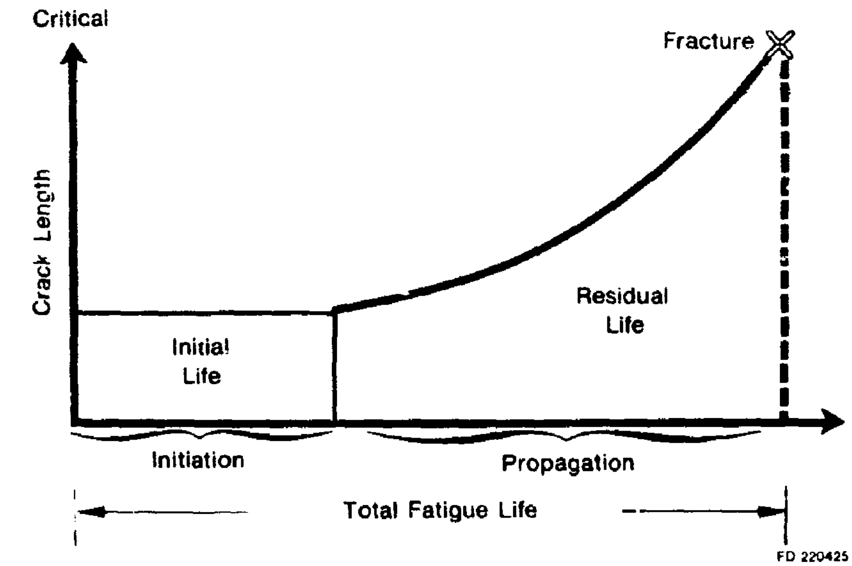

Figure 1 Total Fatigue Life Segmented Into Stages of Crack Development, Subcritical Growth and Final Fracture

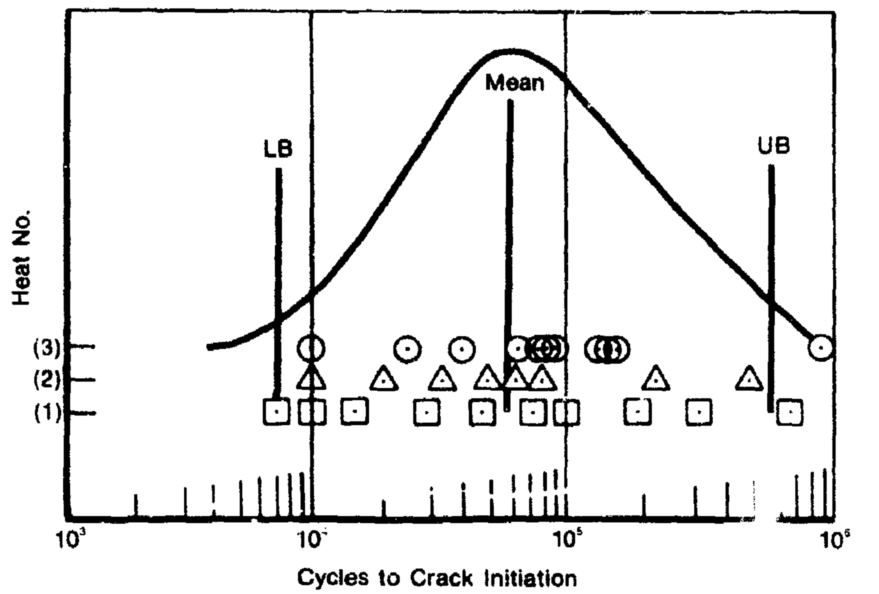

Figure 2 Material Data Scatter Results in Conservative Life Prediction

When considered with other uncertainties in any design system (e.g., stress analysis error, field mission definition, fabrication deviations, temperature profile uncertainty) the final prediction is made for disk crack initiation life for an occurrence rate of 1 in 1000 disks. It is at this life that all LCF-limited disks are removed from service. This procedure has successfully prevented catastrophic in-service failures However, in retiring 1000 disks because one may crack, the remaining life of the 999 good disks is not utilized. The amount of usable life remaining can be significant, as shown in Figure 3. Over 80% of the disks have at least 10 lifetimes remaining.

The means of extracting the remaining useful life from each disk must be safe to avoid catastrophic failure. This is done by determining the disk crack propagation life (N_p) (at every critical location) from a defect barely small enough to be missed during inspection. The RTS interval is then calculated by conducting a life cycle cost (LCC) analysis to determine the most economical safety factor (SF) to apply to the shortest N_p (RTS interval = N_p/SF). Cost vs SF is plotted for each individual disk and combined to determine the most economical interval to return an engine or engine module for inspection. An example is shown in Figure 4.

Ideally, the first required disk inspection is at or near the end of the analytically determined crack initiation life. Only one disk in 1000 of that type inspected should have a crack and be retired. The remaining 999 could be returned to service for the calculated RTS interval. This process is repeated at the end of each RTS interval until a defect is found whose residual life would not allow safe completion of the RTS interval, or conversely, would require such a short RTS interval that it is not economically or logistically feasible. Then, at the end of each RTS interval unsafe disks are retired and the others returned to service. Figure 5 illustrates how the residual life is extracted from each disk after the analytically determined crack initiation life is used.

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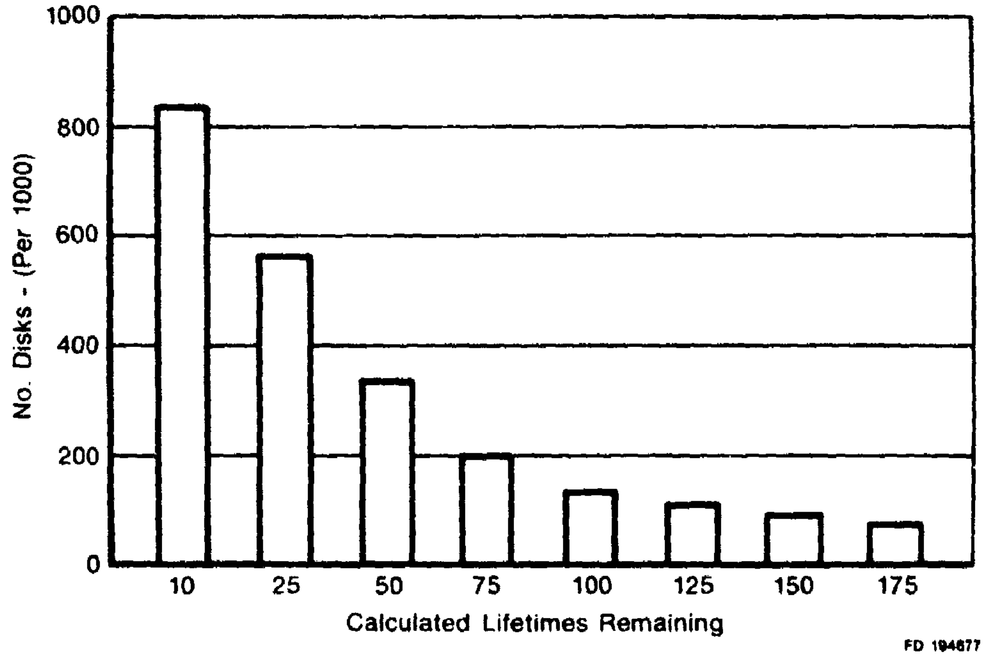

Figure 3. The Majority of Disks Have Useful Life After Retirement

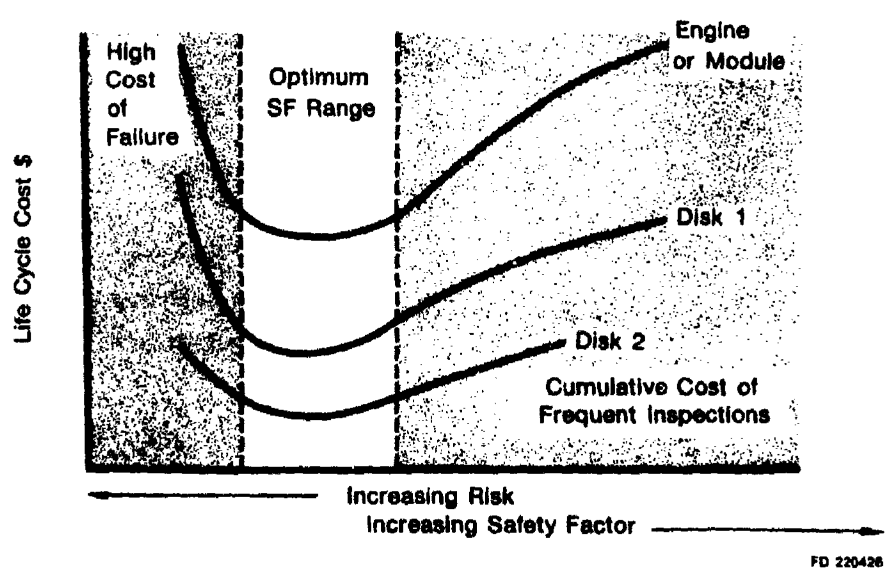

Figure 4. Safety Factor is Determined from an Economic Balance Between High Cost of Failure vs Cumulative Cost of Frequent Inspections

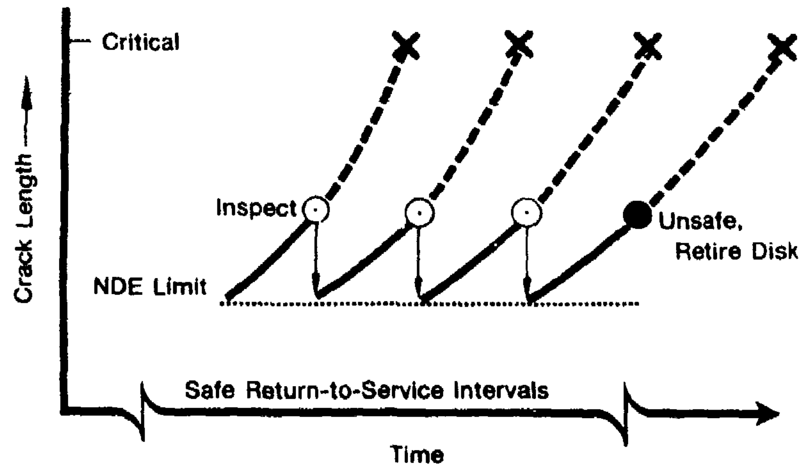

Figure 5. Basic Retirement for Cause Concept

In practice, gas turbine engines are overhauled periodically prior to reaching the crack initiation life limits of their rotor components. This type of procedure could also be used at these overhauls to prevent failure from rogue defects. A RFC maintenance scheme could be tailored to be compatible with, and an extension of, the maintenance "rythm" for a given engine system. Initially, it is anticipated that components selected for RFC would be components that already receive inspections at scheduled overhaul periods. These periods are some increment of the expected crack initiation life; therefore, present methods for predicting crack initiation life are adequate. Because of this, improved accuracy in knowing the initiation life and understanding the initiation phenomena, while desirable, is not absolutely necessary. Also, it is anticipated that upon initial implementation of RFC, any component found to have a service induced defect of any detectable size would be retired *even though a crack propagation analysis would indicate sufficient residual life for additional RTS intervals.* As field service experience builds confidence, or as specific system supportability problems dictate, utilization of the total residual (propagation) life in setting RTS intervals could be done. However, assuming that only the initiation lifetimes shown in Figure 3 were utilized, significant economic benefits would be obtained.

TECHNOLOGY DEVELOPMENT REQUIRED

As Figures 4 and 5 illustrate, RTS intervals are based on two broad technologies: nondestructive evaluation (NDE) and applied fracture mechanics, and evaluated based upon economic factors.

Fracture mechanics must provide an assessment of the behavior of a cracked part should it pass NDE with a defect just below an inspection limit. To assure safe return to service of a part which *may* contain a small crack, an accurate crack propagation prediction is imperative. Recent strides in applied elevated temperature fracture mechanics (References 1, 2, 3, and 4) have provided the necessary mathematical description (models) of basic propagation, i.e., crack growth under conditions of varying loading frequency (ν), stress ratio (R), and temperature (T). Further work (References 5 and 6) has expanded this capability to include loading spectra synergism, i.e., crack growth subjected to (frequent) periodic major load excursions separated by a small number (10-50) of varying subcycles. It is important to note that a typical mission loading spectrum to which gas turbine engines are subjected bears little resemblance to that experienced by air frames, and therefore different predictive tools are required for each (Reference 7).

Referring again to Figure 5, it is seen that accurate propagation predictions constitute a necessary, but not sufficient, condition for the implementation of retirement for cause (RFC). The other requisite technology is high reliability nondestructive evaluation. NDE must provide the means of screening disks with flaws that could cause component failure within an economically feasible RTS interval. Insufficient NDE reliability has been a major argument against implementation of an RFC maintenance program. NDE capability with acceptable flaw detection resolution has been available for some time (References 8 and 9), but adequate reliability of flaw detection has been lacking (Reference 10). Complementary inspections and improvements in NDE single inspection reliability (by automation), can provide the required reliability for many gas turbine engine components to economically utilize the RFC maintenance concept.

The fracture mechanics approach to estimating component service life is based on the assumption that materials may contain intrinsic flaws, and that fatigue failure may occur as a result of progressive growth of one or more of those flaws into a critical crack. Thus, the prediction and monitoring of crack growth as a function of time (or cycles) becomes one of the basic requirements of the analysis system. To utilize such an approach in practice requires quantitative information on component stress, materials characteristics, and nondestructive evaluation capabilities. Much of this information cannot be defined as a single value, but must be described by a probability distribution. Two examples are: the probability that a flaw of a given size will exist in virgin material, or the probability of finding a given flaw size with a standard inspection procedure. In order to obtain a deterministic fracture mechanics life prediction (given these distributions), the conventional approach has been to use worst case assumptions for all parameters. Employing all worst case assumptions (deterministic) necessarily results in a conservative estimate for the service life of the component.

To circumvent this difficulty, the problem can be treated probabilistically. A closed-form solution, which takes into account all the required probabilities, is far too complex to be practicable. An alternative solution is to employ computer simulation techniques. A probabilistic life analysis (References 11 and 12) would use a distribution of flaw sizes. This type of analysis results in failure probability as a function of time, includes NDE reliability, and allows selection of an RTS interval to obtain an acceptable (low) failure probability with realistic NDE reliability.

Both deterministic and probabilistic methods could provide some of the NDE reliability through multiple inspections and/or through higher NDE limits due to shorter RTS intervals. Since many NDE errors are the result of human frailty, multiple inspections and automation can enhance detection reliability. A probabilistic life analysis system, however, would have the ability to accommodate NDE reliability (probability of detection versus crack length) distributions and assess their effect upon RFC efficiency. Obviously, high reliability NDE is desired to optimize the economic benefits of RFC.

THE RFC PROCEDURE

The RFC flow chart (Figure 6) illustrates a simplified view of how this maintenance concept can be utilized. When an engine (or module) is returned for maintenance, an economic analysis is performed on the engine or module (i.e., fan, compressor, high turbine, or low turbine) identified as a participant of the RFC maintenance program. If the module has already been in service for several inspection intervals, the probability of finding cracked parts may be great enough to make reinspection economically undesirable and specific components of that module are retired without being inspected. This is determined by the economic analysis at decision point one and is one of three possible decisions. An unscheduled engine removal (UER) may bring a module out of service that is more economical to return to service for the remainder of its inspection interval than to inspect and recertify it for a new full interval (the second possible decision at point one). The remaining choice at point one is to tear down the module and inspect the parts. During inspection, there again are three possibilities (decision point two). If no crack is found, the part is returned to service. If the disk is found to be unsafe, it is retired. The third choice is to investigate modification or repair of a flawed part. An economically repairable part may be repaired and returned to inspection (decision point three).

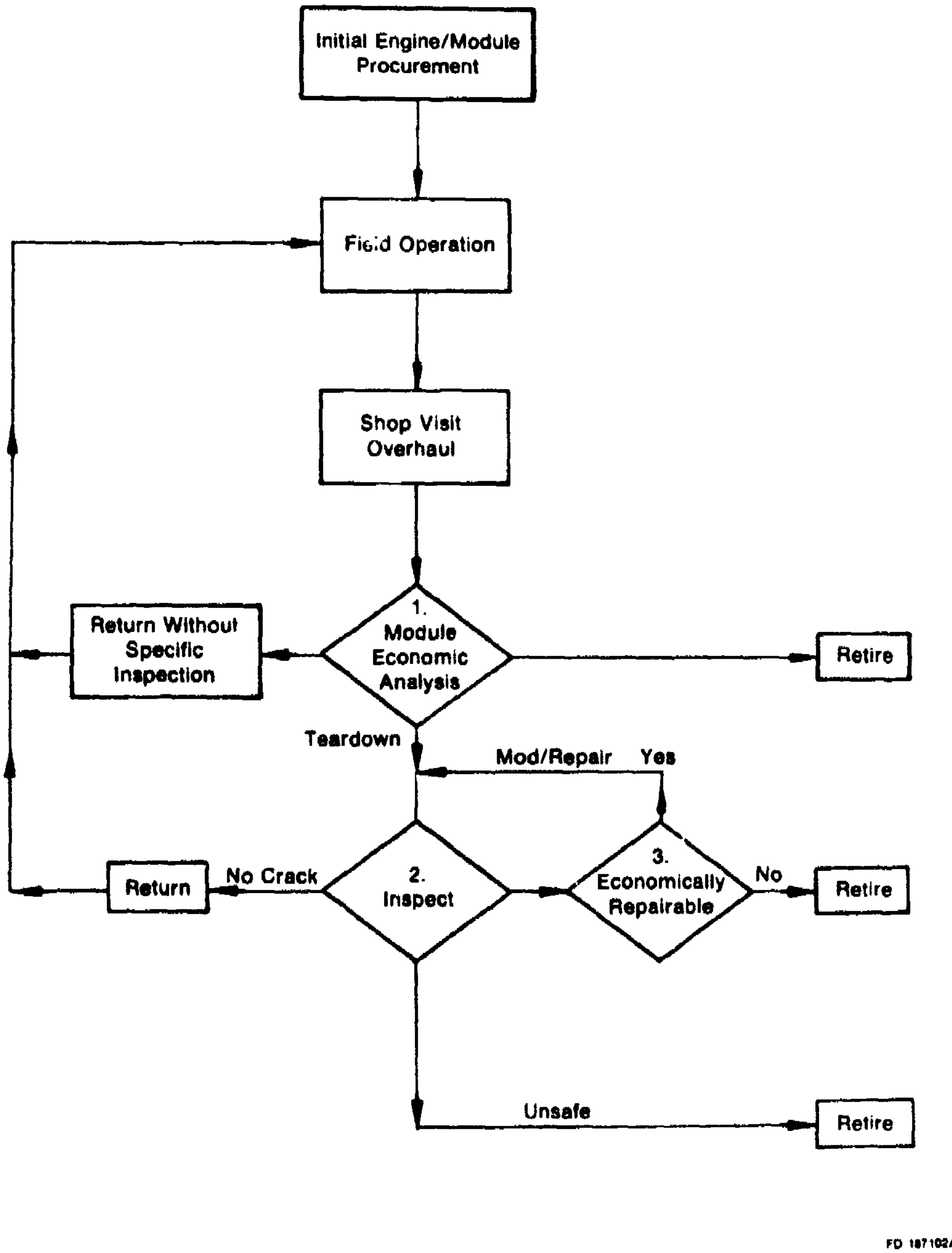

Figure 6. Retirement for Cause Flow Chart

AN APPLICATION OF RETIREMENT FOR CAUSE

The application of a RFC maintenance approach to a military gas turbine engine has been studied (Reference 13), and development activity is underway to reduce the RFC concept to practice. The demonstration and first implementation system is the United States Air Force (USAF) F100 engine. This engine is an augmented turbofan engine in the 110 kN thrust class with a thrust-to-weight ratio in excess of 8 to 1. The engine is currently in operational service around the world in the twin-engine McDonnell-Douglas F-15 and single engine General Dynamics F-16 fighter aircraft.

The F100 is an axial flow, low-bypass, high compression ratio, twin spool engine with an annular combustor and common flow augmentor. It has a three-stage fan driven by a two-stage, low-pressure turbine and a ten-stage compressor driven by a two-stage, high-pressure turbine. The engine consists of five major modules: fan, core (compressor, combustor and compressor-drive turbine), fan drive turbine, augmentor and exhaust nozzle, and gearbox. Each module is completely interchangeable from engine-to-engine at the intermediate maintenance level. The modular approach was selected so that either functionally or physically related parts can be removed as units.

The objective of the study program was to determine the feasibility of applying a RFC maintenance approach to the USAF F100 engine. The study was directed primarily toward rotating components of that engine, specifically the various fan, compressor and turbine disks and the spacers/air seals that comprise the prime rotor structure. The effort addressed the following five major areas:

- Definition of an RFC methodology
- Evaluation of the disks and other appropriate engine rotor components for RFC applicability
- Assessment of the nondestructive evaluation requirements for implementation
- Establishment of a component ranking for development priorities
- Establishment of development plans leading to implementation.

The methodology has been discussed in the preceding section of this paper. The life analysis of every major component was reviewed using deterministic lifeing procedures. Based upon this life analysis, individual component cost, and a 15-year average engine system life, 21-candidate rotor components were selected. The components include: the three fan disks, four of the compressor disks, the four turbine disks, and ten airseals/spacers from the fan, compressor and turbine. These components are all produced from wrought titanium and nickel base alloys.

Each critical area of each component was then analyzed for residual life to establish the type of defect and the detection requirements range for optimal economic benefit. A composite sketch of typical rotor components is shown in Figure 7. As can be seen, the configuration of these components is complex, and innovative techniques are required to enable reliable inspection. The types of defects to be detected include: corner, surface, through-thickness, and internal flaws. Techniques currently exist, such as eddy current, ultra sound, penetrants or proof tests, with the ability to detect these flaws in the sizes anticipated. The major problem to be solved is detecting these flaws in a high-volume maintenance environment with sufficient reliability to enable RFC to be economically viable.

The ranking of components, or groups of components, was based upon three factors: life cycle cost impact, NDE requirements, and an assessment of where the state of the art in applied fracture mechanics is at this time. The final ranking of components for the F100 was done by engine module because NDE requirements are similar among components of the same module; it is economically and logistically impractical to return units of less than a complete module to an engine overhaul center. The development priorities are: compressor components, high-pressure turbine components, low-pressure turbine components and fan components. This order was chosen based upon realistically meeting the NDE goals and the life cycle cost savings per module. In all probability, the chronological order in which RFC would be applied to this engine is: fan, low-pressure turbine, compressor, and high-pressure turbine based upon present technology.

From the results of the first four activities, a preliminary plan was developed to identify the technology and other activities required to enable implementation and operation of RFC at an engine overhaul center, and the time phasing necessary for a target application date of 1985. The logic upon which the plan was developed is shown in simplified form in Figure 8. There are five sequential steps inherent in the planning process: (1) development of the required technological and management tools, (2) establishment of inputs to and exercising the tools, (3) demonstrating the tools, (4) evaluation and documentation of the tools, and (5) implementation and use of the tools. The resulting development plan is shown schematically in Figure 9 and addresses the technology developments discussed in previous sections of this paper. At the present time, these activities are underway and proceeding according to the planned schedule.

Benefits of Retirement for Cause — The assessment of the benefits of a Retirement for Cause maintenance approach to a gas turbine engine is contingent upon many assumptions. These assumptions include: fleet size, anticipated usage rates, usage life, inspection interval, labor costs, parts cost and many others.

To quantify the benefits of this maintenance concept for the USAF F100 engine, life cycle cost analyses were conducted. These analyses determined the change in life cycle costs of the F100 engine that could accrue based upon implementation of an RFC maintenance procedure in 1985 as opposed to a continuation of current or baseline maintenance practices.

The life cycle cost benefits amount to an approximate U.S. (1979) $250 million savings over a 15-year period. In comparison to the investment required, the development and implementation of RFC are extremely attractive.

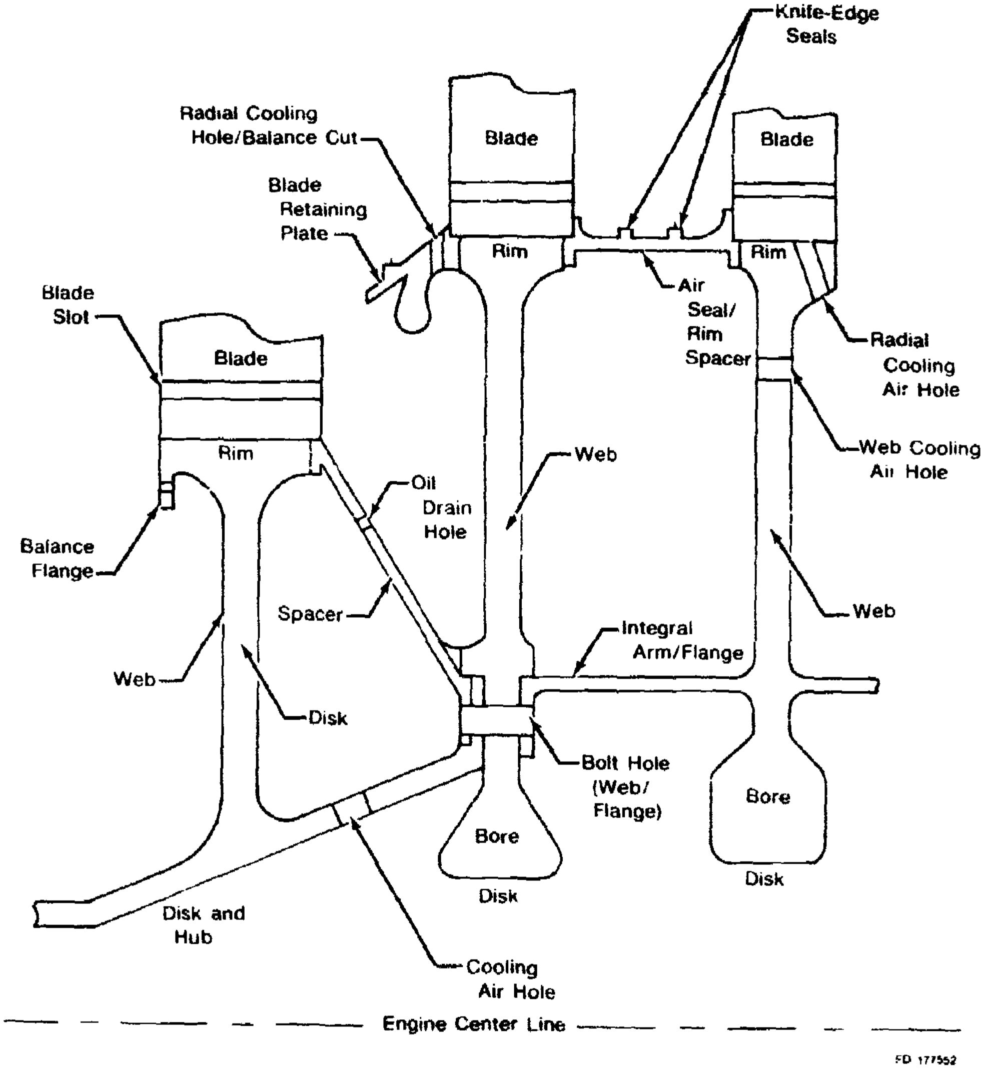

Figure 7. Composite Sketch of Typical F100 Rotor Components and Flaw Types (Not All Features on All Parts)

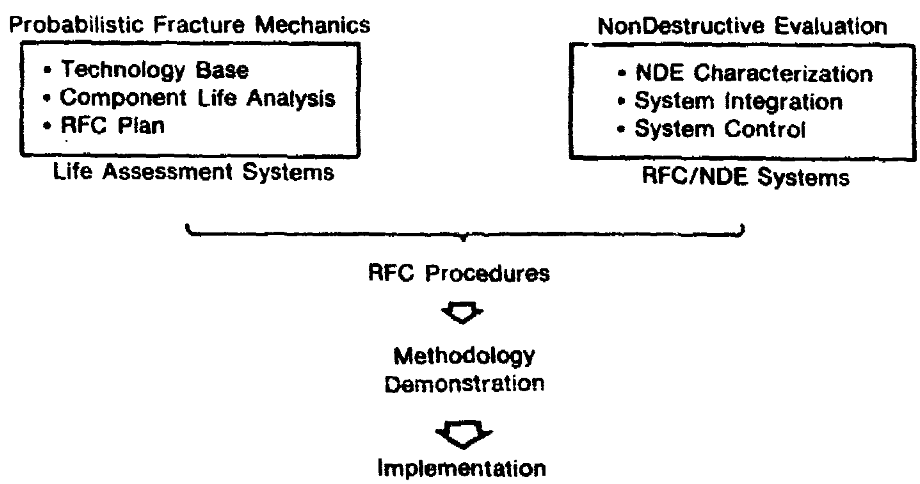

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Figure 8. Technology Development Required for Engine Component Retirement for Cause

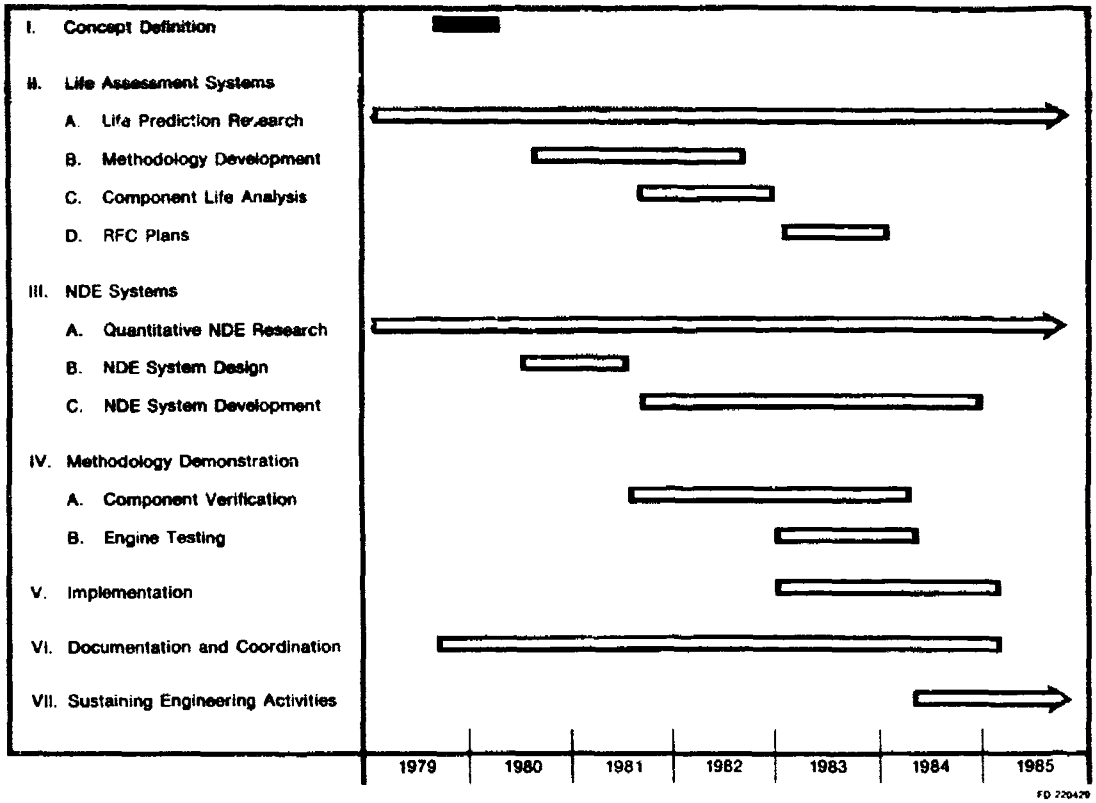

Figure 9. Development Plan for Engine Component Retirement for Cause

CONCLUSIONS

Realization and implementation of a Retirement for Cause Maintenance Methodology will result in system cost savings of two types: direct cost savings resulting from utilization of parts which would be retired and consequently require replacement by new parts; and indirect cost savings resulting from reduction in use of strategic materials, reduction in energy requirements to process new parts, and mitigation of future inflationary pressure on cost of new parts. With the worldwide concern over the availability of strategic materials, the resultant reduction in requirements for, and conservation of critical resources may eventually become as large a factor in applying RFC to specific systems as the direct cost savings upon which the decisions are currently based.

The methodology and procedures described herein are applicable to systems other then the F100 engine. A cursory review of other gas turbine engines indicates that the RFC maintenance concept is generic and has direct applicability to rotor components of those engines. In fact, the methodology has broad applicability to other types of engine components, and indeed, to systems other than aircraft gas turbine engines. The decision to apply RFC to other components or systems would be based upon economic factors, predicated upon the remaining anticipated service life of that system, and it could be a viable maintenance concept for life limited components of all types.

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The concept of retirement for cause (RFC) as presented in this paper has evolved based upon the work of many individuals within the industrial, academic and government communities. The authors acknowledge these contributions which have enabled RFC to move towards reality as a maintenance procedure for military gas turbine engines. Special acknowledgement is accorded to Dr. W. H. Reimann, Air Force Wright Aeronautical Laboratories, Materials Laboratory, Wright-Patterson Air Force Base, Ohio, for his technical contributions, management sponsorship and perseverance in guiding RFC of gas turbine engine rotor components from idea to reality, and to the Defense Advanced Research Projects Agency, Washington, DC for their sponsorship of the contract program activities.

DEFECTS AND THEIR EFFECT ON THE BEHAVIOUR OF GAS TURBINE DISCS

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Summary

All materials contain defects but recent moves to higher stress levels have lead to an increasing number of these defects being malignant. Unless the method used to life fatigue critical components like gas turbine discs allows assessment of defect presence and behaviour, the risk of serious failure is dramatically increased.

A method is discussed where defect behaviour can be assessed as part of a total life approach to disc behaviour prediction and is explained together with the effects of differing defect types. Such an approach gives realistic manufacturing standards and controls and leads directly to an 'on condition life' approach.

1. Introduction

Since the realisation, some 30 - 40 years ago, that gas turbine discs were the most critical components in the engine because pieces cannot be contained if failure occurs, various criteria have been used to define both the design parameters and lives of such components. Until the end of the 1950's the main factors considered were creep, for the rims of turbine discs, and tensile strength, to give a satisfactory overspeed margin without burst, for both compressor and turbine discs.

The lessons were learned early - as can be seen from the failed turbine disc in the Whittle Mk3 engine in the British Science Museum (Figure 1).

In the period up to the mid 1960's failures were due, in the main, to lack of quality control, where the material didn't meet specification, to design problems mainly at stress concentrations (dealt with by geometry modification) or to problems such as corrosion and fretting which were alleviated by material change or coatings. In Britishturbines in particular, little attention had to be paid to low cycle fatigue problems because of the extensive use of martensitic steels where the limitations for overspeed meant a fatigue life well beyond that practical in service.

This situation changed with the introduction, first of nickel base and then titanium alloys in both turbine and compressor discs where, although creep and tensile criteria still needed observing, it was found that the most important factor governing component design and performance was low cycle fatigue.

The literature published on the subject of fatigue fills many library shelves, but, as applied to gas turbine discs, this approach assumes that materials initiate failure origins that turn into cracks in a way that can be predicted from relatively simple laboratory tests, with a variation in performance that can be handled statistically. This method of life prediction, which is still used in many areas of the industry, is based upon assumptions that the bounds of the material behaviour can be circumscribed by manufacturing control and the presence of defects eliminated by inspection (Figure 2).

This paper sets out to show that recent developments both in materials and design now require a more complex approach to disc performance and associated manufacturing criteria.

2. Basic Lifing Approach

The licencing authorities for civil engines demand that the manufacturer establishes the finite life of some 'best' specimen and then allows 'in flight' use of a proportion of this life to cover the 'worst' expected example of the disc family. A similar lifing system is used for military discs. The exact way of defining the finite life varies from authority to authority (and the mode of establishing that life from manufacturer to manufacturer).

The British authority (CAA) and Rolls-Royce have always used cyclic spin pit testing under near engine conditions as the basis of disc lifing. Whilst laboratory materials fatigue data is used in the design context it has not been relied on to calculate in-service performance.

Such an approach is based on careful control of total manufacture to make sure that the components tested are in the same family as those in service, together with careful control of inspection. The end point in life has, in the main, been taken as ⅔ burst life rather than 'first crack'. The latter is rather a nebulous concept when regarded as a life criterion (Figure 3).

In reality this has lead to a lifing approach which has depended on all three stages of life - 'initiation', propagation' and 'final failure'. At first this was, of course, empirical but during the past 10 years the materials engineer has had techniques and understanding to control the various life phases and take account of the increasing temperatures and stresses the designer is imposing on materials. Detailed examination of the many rig-failed parts required to support this lifing system has given some insight into the behaviour of materials under disc imposed conditions and also the interaction of life with the material structures and defects that occur in reality.

3. Stages in Disc Life

The life of the component can be considered as four stages :-

3.1 Crack nucleation
3.2 Growth to material grain size
3.3 Stress dominated crack growth
3.4 Final failure

3.1 and 3.2 are usually considered together under the 'Initiation' concept. The end point of 'first crack' considered by many designers can occur in either stage 3.2 or 3.3 depending upon definition of size of first crack and material grain size.

3.1 Crack nucleation is the process of accumulation of fatigue damage in the material until a physical discontinuity or 'crack' occurs. This can happen in a number of different ways, both at the surface and within the component, whether non-coherent defects are present or not.

As materials of higher strength are used they tend to strain localisation types of behaviour in this stage which increase the propensity for sub-surface failure (Figure 4). The life of this phase depends upon imposed stress/strain (including presence or not of defects) temperature, orientation.

3.2 The crack then grows to material grain size. This type of growth is governed by the same factors as 3.1 - that is local conditions of stress and strain at the crack which are those of the individual grain that cracked. The life in this phase can vary from nil for some materials where the nucleation process is one that produces grain size cracks to many times that of the nucleation phase where growth at this stage involves a series of nucleation process - each at higher local stress until the crack propagation threshold is reached.

3.3 Once the crack can be considered large i.e. above grain size, its behaviour is dominated by the imposed stress system in the mode normally predicted by the crack propagation phase of fracture mechanics analysis. Outside temperature and stress such behaviour is independent of most parameters except basic alloy system - most martensitic steels behave the same, as do nickel and titanium alloys but each group has significantly different crack propagation rates from the others.

3.4 This crack grows until some final failure criterion is exceeded. This normally is dictated by fracture toughness criteria, the outset of unstable crack growth, but can also be that the section stress exceeds the tensile level or that some sort of plastic tearing mode occurs. A variety of failure prediction concepts are available but final size at burst has little effect on overall life - although it obviously has inspection implications.

The percentage of life spent in each of these phases varies according to material, stress/strain level and presence of defects.

4. Defects

Defects can be classified into two types :-

4.1 - Physical discontinuities
4.2 - Micro structural deviations that promote scatter in behaviour outside that designed for.

4.1 Physical discontinuities

This category covers cracks, inclusions, areas of contamination, slag and other non-coherent areas within the component.

They affect component life by either eliminating or severely reducing the 'initiation' or crack nucleation phase. The effects are geometrical; that is, such defects create a disturbance in the section stress field which accelerates the build up of fatigue damage, leading to early crack formation.

If their presence and geometry is known then their effect can be accounted for in the life method, either empirically by disc testing or theoretically by stress calculation and laboratory testing.

4.2 Structure deviations

This category covers coherent defects where any volume within the main body of the material behaves differently under the imposed stress conditions. The defined scatter of material properties depend upon defined bands of microstructure and chemistry.

In volumes of material where the latter factors lie outside the defined limits the material has a different stress-strain-time behaviour which imposes higher than predicted stresses in or about the defective area leading to more rapid failure.

Both types of defect are, and always have been, present in even the best commercially manufactured components. Their presence can, however, either be benign or malignant, depending upon their size and the imposed stress conditions. The traditional 'no defect' approach has relied on eliminating all malignant defects. Current materials and operating conditions are making such an approach unrealistic as defects below NDI method detection limits are in some cases now malignant.

5. Methods of accounting for defect behaviour

Disc lifing requires an understanding of component life in the terms described in section 3 above so that appropriate data and methods can be applied to each stage. The effect of various types of defect can then easily be assessed provided their presence is recognised and their behaviour known - at least within bounds, if not exactly. This approach can then be applied to develop a first life for an individual component using modifications of existing fatigue life and fracture mechanics methods, and to subsequently extend component life in further increments using some form of inspection - a life on condition or retirement for cause system.

This overall assessment of life also allows the practical considerations of such factors as frettage, corrosion and mechanical damage.

The type of data and approach required for each life stage is discussed below.

5.1 Crack nucleation

Knowledge of the nucleation mechanism is all important here - in particular is this likely to be associated with a non-coherent defect or does the crack nucleate at some coherent structure origin?

In the former case life declaration for this phase depends upon definition of defect size and knowledge of the conditions when the defects become malignant. If they are benign e.g. silicides in Titanium alloys, then their presence can, of course, be ignored. Extrapolation of previous experience to higher stress levels can be very misleading in such cases.

Definition of defect size depends upon practical knowledge and experience of various NDI methods and their interpretation to reality. Figure 5 shows the relationship of true size and ultrasonic response for a certain component in Waspaloy. In particular it shows the statistical nature of this type of information - which also applies to eddy current, penetrant or binocular inspection of surface cracks.

Knowledge of defect type depends upon detailed understanding of the total manufacturing sequence so that the chance of getting a certain type of defect in a particular volume of the disc can also be taken into account - another statistical factor.

Understanding of the defect behaviour usually requires first hand observations of behaviour of components containing defects at representative stress/time/temperature conditions. Sometimes this shows that the geometrical effect can be assessed by 3D finite element techniques (Figure 6) but more often the onset of defect malignancy and its extent can only be determined empirically. Even where many defects are present - as in powder nickel base discs - their behaviour can only be statistically assessed by interpolation to engine conditions from laboratory and component tests.

Where malignant defects are not present and crack nucleation occurs at structural features it is important to both understand and control the manufacturing process so that the variation of nucleation sites present lies within the limits required to control the scatter band of behaviour to that assumed in the life calculations.

Again the approach requires a detailed knowledge of the mechanism, and interpolation of data. Extrapolation of experience to higher stress levels can be misleading! A number of cases have been recorded where a modest change in structure has promoted an entirely new mode of behaviour (Figure 7).

5.2 Small Crack Growth

Whilst this phase, if it exists at all, must be considered separately from 5.1 most of the above remarks on defect dependence apply. The most important factor is nucleated crack size and whether, under the locally applied operating conditions, it is capable of propagation or requires a number of renucleation steps - each time from a bigger 'defect'.

As the crack is small its behaviour is dominated by local conditions within the material and so assessment of length of this phase must be statistical.

5.3 Large Crack Growth

This phase of life is probably the easiest to assess as the fracture mechanics techniques required are now well known and the scatter of behaviour small. The main factor of uncertainty is in definition of the defect size at the start of the phase - particularly if a sub-surface site is involved.

5.4 Final Fracture

By the time the crack has reached the size for final fracture, growth is so rapid that total life is relatively insensitive to this factor. It is important, however, in assessing the risk of in-service failure which is higher if the final fracture size is similar to the limit of inspection method available.

Special attention has to be applied where change in temperature promotes a step in toughness for steels or causes other such discontinuities in behaviour.

The statistical nature of all these phases must be noted and the life declaration for the component can only be expressed in a statistical form - a risk analysis for any given calculated life should be declared (Figure 8). This involves defining some definite life end point - like ½ burst life rather than an ill-defined 'first crack criterion' or a quasi-quantitative crack size e.g. .75mm which gives a dramatically different end point depending upon the stress conditions and material involved - in addition to the problem of coping with 1mm defects using that approach!

Whilst the above has concentrated on material behaviour factors it also has to be emphasised that the approach requires excellent 3D stressing of the component including detailed knowledge of thermal and residual stresses. Without this knowledge of stress levels and gradients - which in itself depends upon detailed stress strain knowledge of the material - the effect of all important component geometry can nullify other considerations.

6. Case studies of 'defect' behaviour

The application of this approach to a number of real cases of defect presence and behaviour is discussed briefly below.

6.1 Failure of a large Titanium Fan Disc

Conventional lifing of a fan disc had defined a life of more than 10,000 flights with final fracture occurring at the disc bore. Whilst spin pit testing confirmed a high life, final failure occurred from the back face. As investigations were proceeding to examine this discrepancy two in-service failures occurred (Figure 7) in less than 1,000 flights.

The lifing method and predictions were based on laboratory specimen testing and spin pit testing of typical samples of the appropriate alloy at 15 to 20 stress cycles/minute. Investigations of the problem (which was difficult as the in-service failures were not recovered in either case) showed that the effect of three unconsidered and unknown parameters had combined to cause early failure. These were :-

a) The manufacturing route of the disc lead to high residual tensile stresses at the back face which moved the most critically stressed position away from that predicted.

b) Ingot porosity occurring in the ingot hot-top region was not healed by subsequent metal working and was too fine to be detected by available NDI techniques

c) The material microstructure in the back face region was a coarser, slow cooled version of that of the typical alloy, which promoted a step change in established behaviour. The material at the failure site showed alignment over significant areas which allowed the small pores to rapidly become large propagating cracks and these cracks to propagate at two orders of magnitude faster under service 'dwell' conditions than that measured under normal test conditions. The net effect is shown in Figure 9. Application of the type of lifing approach shown above allowed the problem to be identified and reproduced even though the failures weren't recovered, and solutions to be found. These showed that a safe, albeit short, life could be tolerated in the immediate period whilst a long term solution was reached. It also showed that a material change was required and why.

6.2 Behaviour of a Waspaloy Turbine Disc

A rig test disc, being used for normal life development showed extensive early cracking from a sub-surface defect well inside the ultrasonic acceptance level (1.25mm flat bottom hole equivalent).

The defect behaved predictably once it was assumed to be a crack (Figure 10). Subsequent investigation established that it had given a small ultrasonic response (around .7mm fbh) even though it was large and malignant (Figure 11). To prevent in-service failures a new ultrasonic standard was adopted related to real defect size and the component life related to the largest undetected defect size - the life method detailed above was applied.

6.3 Use of powder Ni alloys as engine discs

Consolidated Ni-base powder discs inherently contain defects from the manufacturing process. These defects are below the detectability of current NDI methods and yet they are the factor governing component behaviour (Figure 12).

A comparison of lifing methods and acceptable parameters is shown in Figure 13 where the 'defect' governed behaviour is presented in a fatigue curve form for comparison with a traditional fatigue approach. Application of the total life approach shows that the use of such a product can be safe provided defect behaviour criteria are used; that the extrapolation of traditional fatigue and tensile design criteria to higher stresses can be very misleading, causing early component replacement at best - in-service failure at worst; and that material development needs to account for defect tolerance when the alloy/structure compromise is reached.

7. Implications of Total Life system (Figure 14)

Rolls-Royce have been applying this approach of disc lifing in an increasingly comprehensive way over the past eight or nine years - based on extensive laboratory materials testing, component spin pit testing and manufacturing defect investigation. Experience has revealed the following advantages:-

a) Identification of quantitative defect standards for all features (including welds) which allow realistic definition of manufacturing allowable against component operating requirements

b) 'Hazard review' of manufacturing processes to identify critical phases allowing a safe but economic approach to product specifications on the basis of operating requirements.

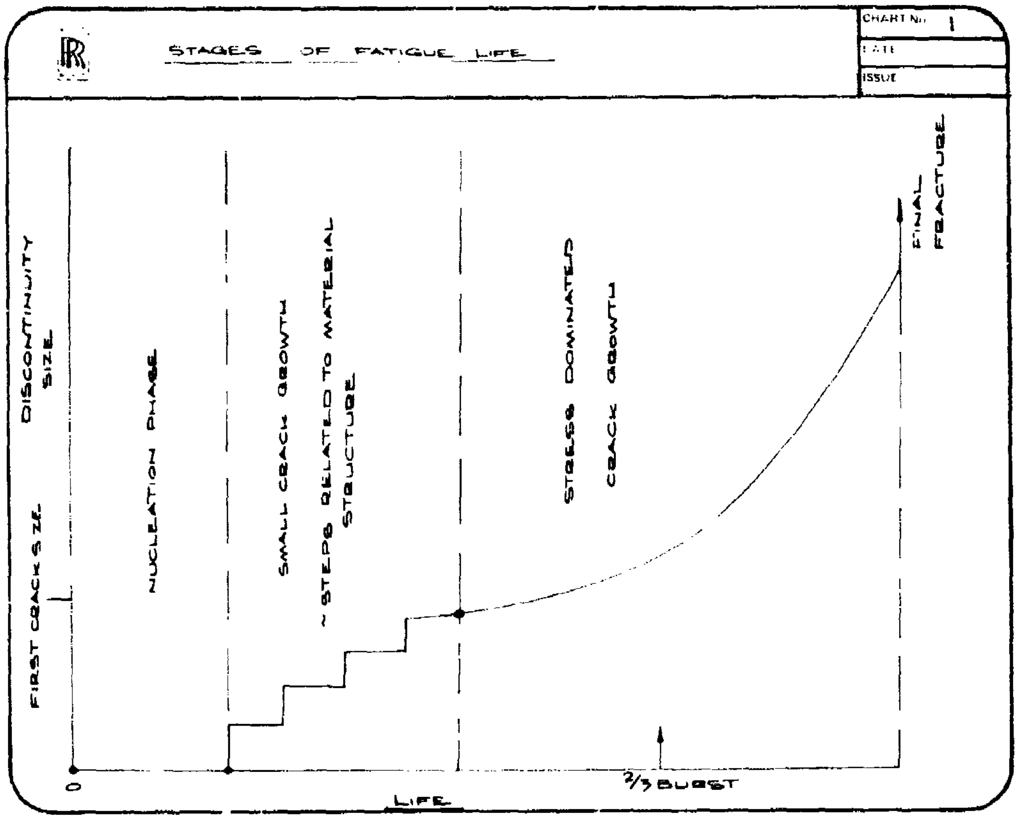
c) Safer use of material at high operating stress levels with a realistic approach to design allowables.

It is already allowing a 'life on condition' approach to individual component life extension in service beyond the initial 'family life' as well as realistic use and definition of inherently defect-limited materials and processes such as powder nickel, castings and welds. Whilst the spread of life on condition applications are being held back by NDI capability the total life method still allows a total examination of component manufacture and operation to achieve a safe economic life - the only way that the high strength materials required by the new efficient aero dynamic designs can be used.

8. Acknowledgements

The author acknowledges the help of his friends and colleagues - particularly Mr A C Pickard, for the mysteries of 3D finite elements and Mr D Gibbons for practical lifing assessments.

Professor D W Hoeppner also provided inspiration through the difficult times.



METHODS FOR EACH LIFE PHASE

CHART No. 2

DATE

ISSUE

NUCLEATION	SMALL CRACK GROWTH	STRESS DOMINATED CRACK GROWTH	FAILURE
MATERIAL FAILURE	CRACK PROP.	FRACTURE	CRITERIA
MECHANISM WITH	THRESHOLD	MECHANICS	E.G. RAPID
APPROPRIATE STRESS	DATA RELATED	DATA AND APPROPRIATE	FRACTURE
/LIFE DATA.	TO STRUCTURE	STRESS INTENSITY	C.O.D.
NUCLEATED	STRUCTURE	FACTORS.	TENSILE
DISCONTINUITY SIZE.	DOMINATED CRACK	INITIAL DEFECT	BUCKLING.
PRESENCE OF	GROWTH MECHANISMS	SIZE.	
MALIGNANT DEFECTS	& RATE.		
POSSIBILITY OF	ONSET OF		
EXTRANEOUS EFFECTS	STRESS DOMINATED		
– FRETTAGE, CORROSION	GROWTH.		
MECHANICAL DAMAGE.			

CHART No 3
DATE
ISSUE

DEFECT TYPES

PHYSICAL DISCONTINUITIES

- CRACKS
- INCLUSIONS
- SLAG ENTRAPMENT
- VOIDS & POROSITY
- DENTS & SCRATCHES
- FORGING LAPS & FOLDS

STRUCTURE DEVIATIONS

- CHEMICAL SEGREGATION
- METALLIC CONTAMINATION
- UNCONTROLLED HEAT TREATMENT
- GRAIN BOUNDARIES SPECIES
  - ~ CARBIDES IN Ni BASE ALLOYS
  - ~ α IN Ti ALLOYS.
- GASEOUS CONTAMINATION

CHART No 4
DATE
ISSUE

IMPLICATIONS OF TOTAL LIFE METHOD

a.) IDENTIFICATION OF REALISTIC DESIGN AND MANUFACTURING ALLOWABLES.

b.) REALISTIC PRODUCT SPECIFICATIONS BASED ON COMPONENT OPERATING NEEDS.

c.) SAFE USE OF MATERIAL UNDER HIGH STRESSES AND OTHER EXTREMES OF POSSIBLE OPERATING ENVELOPE.

d.) ASSESSMENT OF EFFECT OF EXTERNAL FACTORS E.G. CORROSION & FRETTAGE.

e.) INCORPORATES LIFE ON CONDITION OR RETIREMENT FOR CAUSE APPROVALS.

Fig.1 - Failed Turbine Disc - Whittle Mk 3 Engine (Courtesy Proc. Inst. Mech. Eng. 1945 152 419)

Engineering premise

All materials are homogeneous, elastic, isotropic media, free from defects and amenable to conventional fatigue cumulative damage analysis within normal scatter limits

Fig.2 - The 'Engineers Premise'

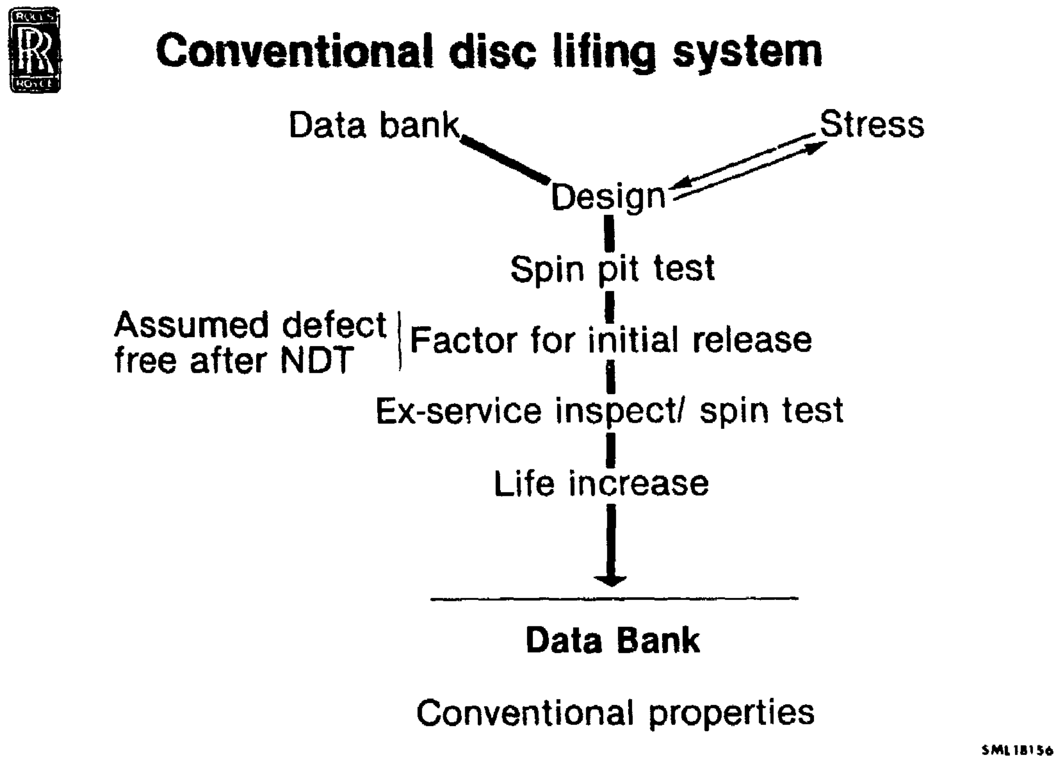


Fig.3 - Conventional Lifing System

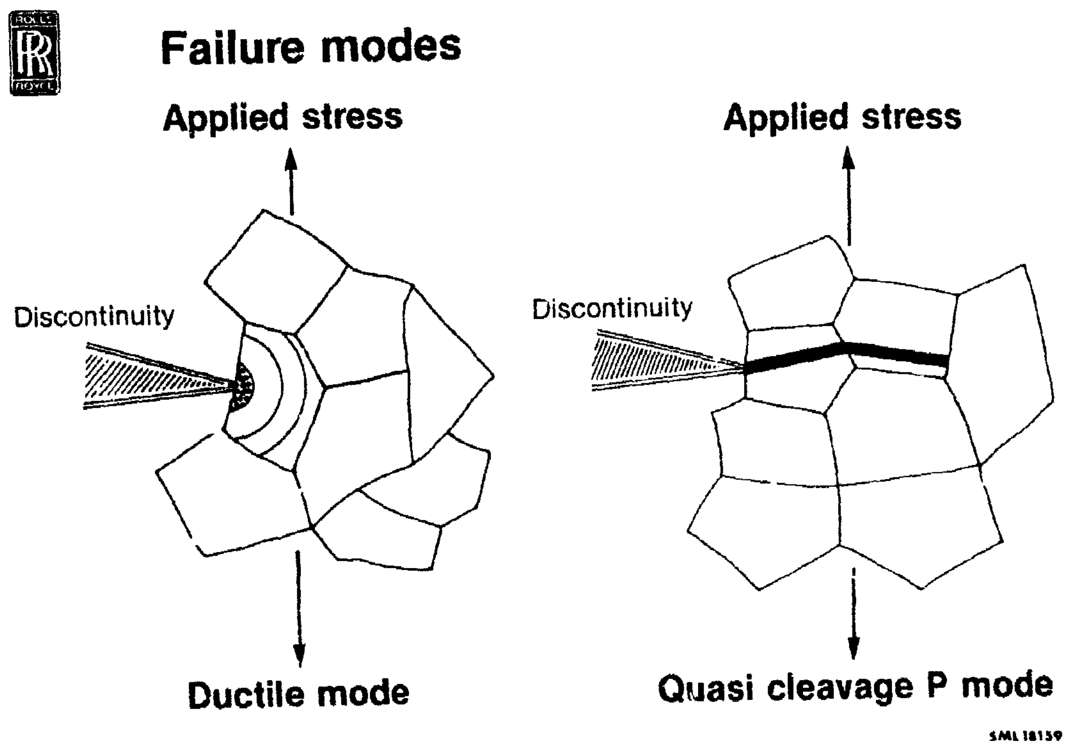


Fig.4 - Crack Tip Failure Modes

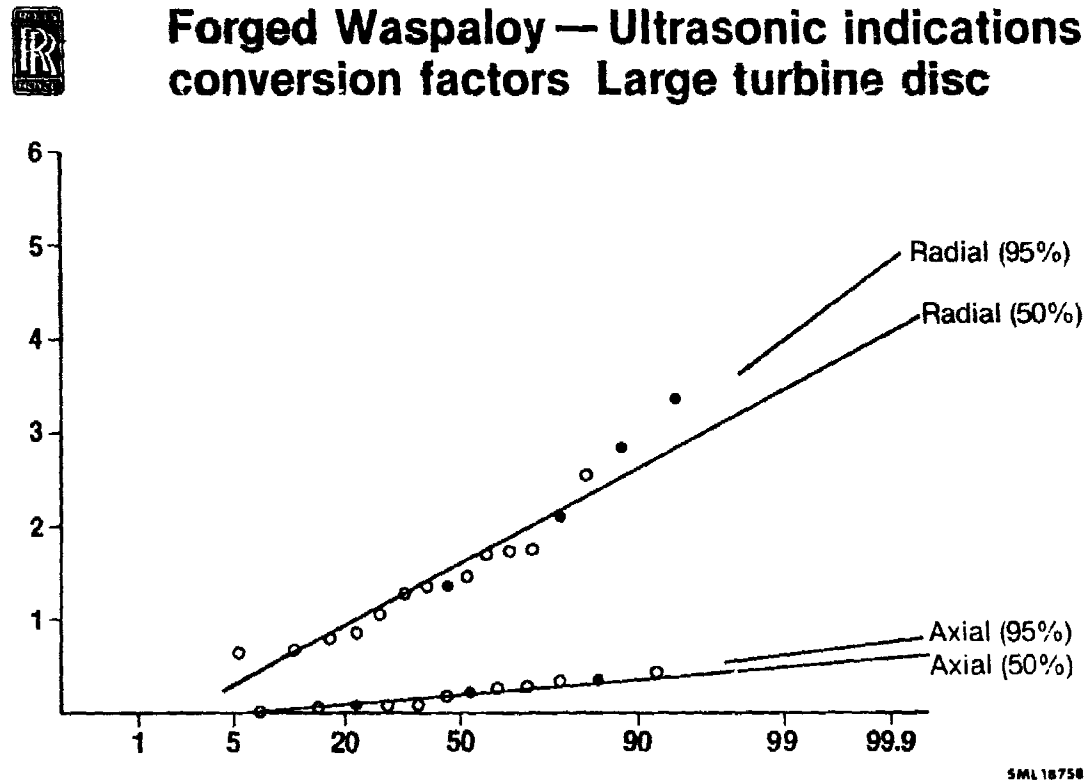


Fig.5 - Ultrasonic Response of Defects

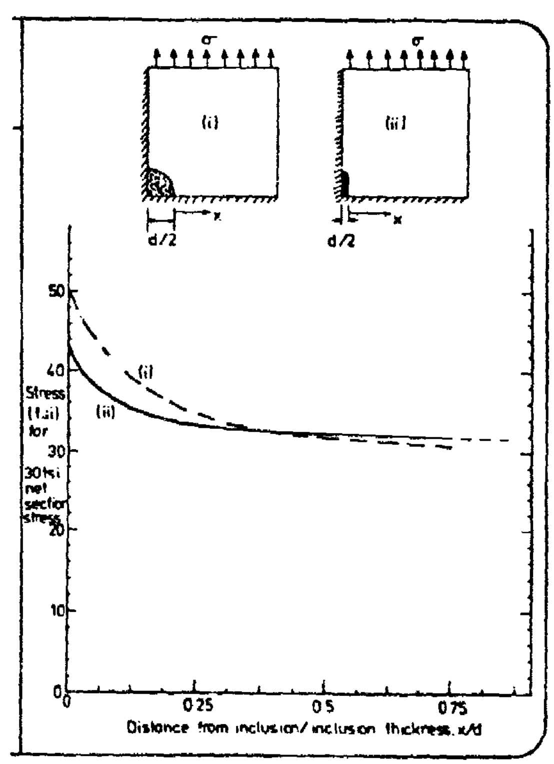


Fig.6 - Defect Stress Effects

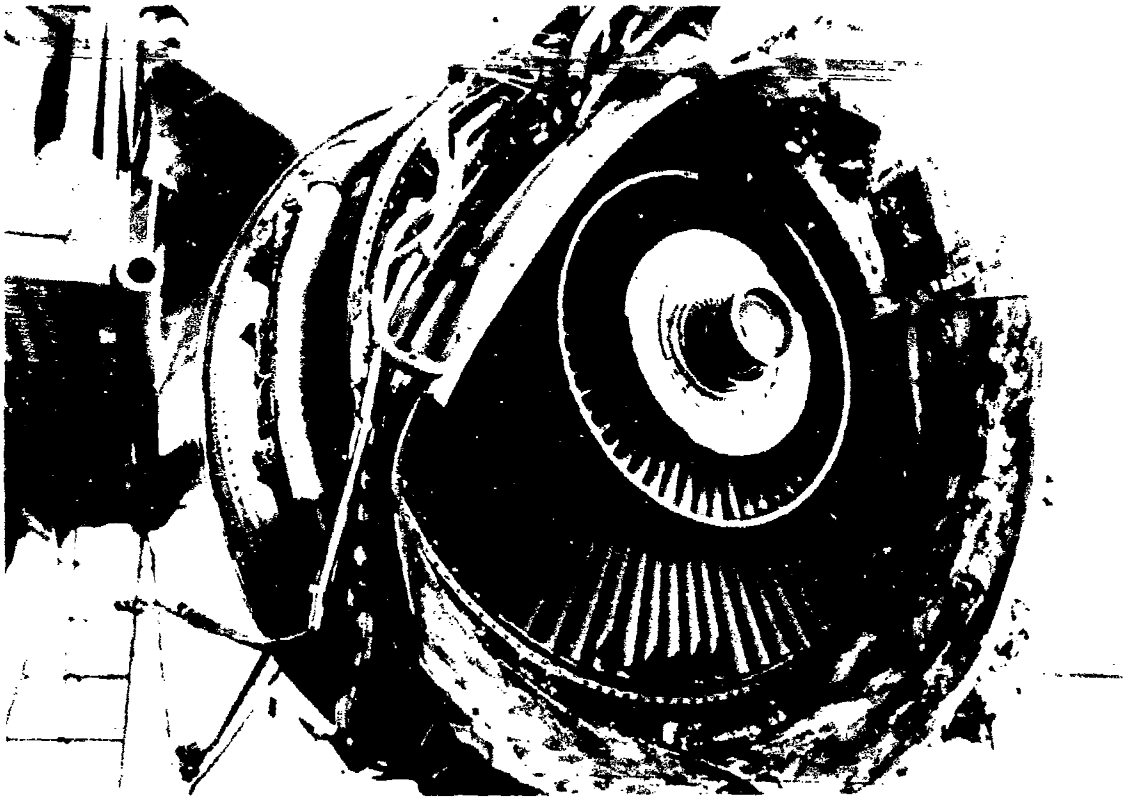


Fig.7 - Effects of a Fan Disc Failure

Risk analysis

- Probability of the component containing a defect
- Probability of the defect being detected
- Probability of the defect being malignant
- Probability of the defect being in a particular position and orientation in the component
- Relationship of the NDE indication to the real defect size - cut-ups of components containing defects
- Probability of a defect in a given position being large enough to cause failure in a given number of cycles or flights
- Variability of material properties
- Probability of a failure being hazardous
- Definition of an acceptable probability of failure (or hazardous failure)

Fig.8 - Risk Analysis

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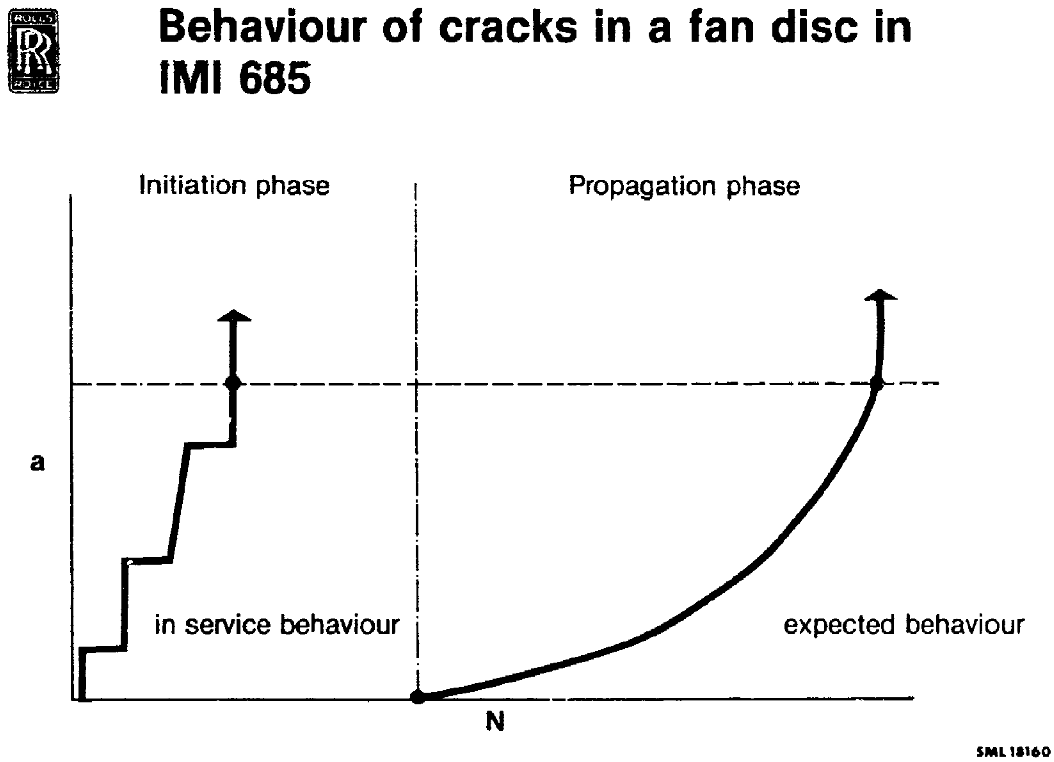


Fig.9 - Crack Progression in a Fan Disc

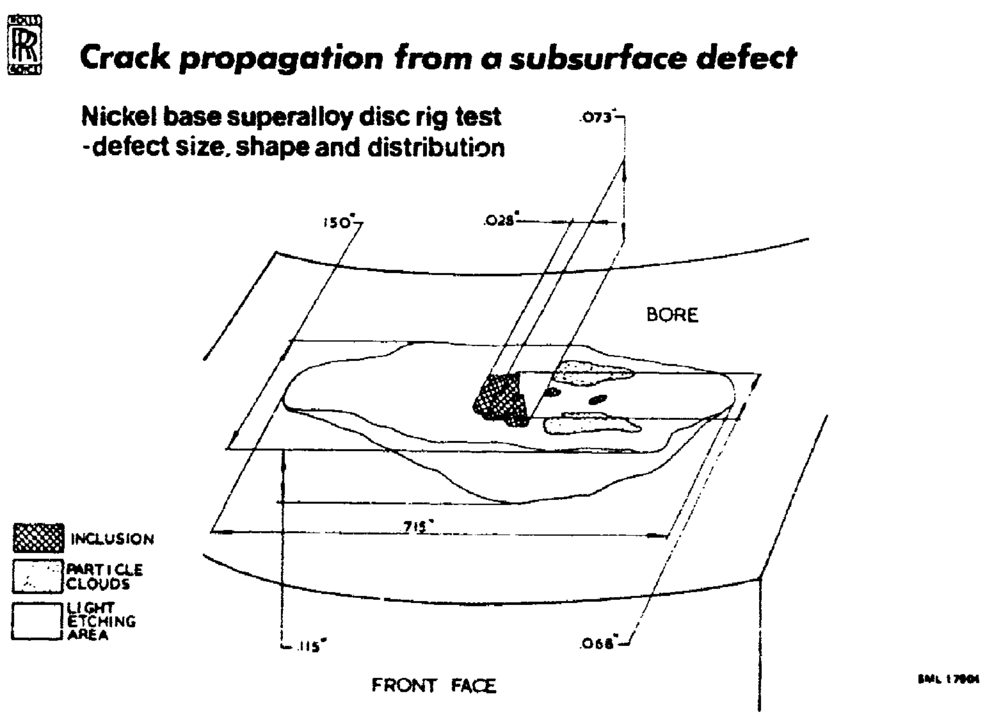
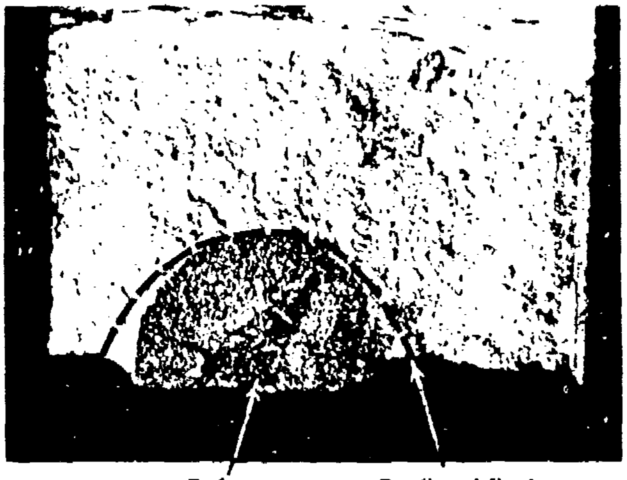


Fig.10 - Propagation of a Sub-Surface Defect

Crack propagation from a subsurface defect nickel base superalloy disc rig test



During manufacture, a defect was detected in the bore using ultrasonics, at -11dB to -12dB level

The disc was rig tested until failure occurred in the rim region. The bore crack shown was then opened.

Ratio of predicted to actual life, using CT specimen data, is 0.86.

Ratio of predicted to actual life, using corner crack data, is 1.07

Ratio of predicted to actual life, from striation count is 1.44 ± 0.48.

Fig.11 - Sub-Surface Defect in Waspaloy

POWDER ASTROLOY

Low Cycle Fatigue Failure from Subsurface Defect

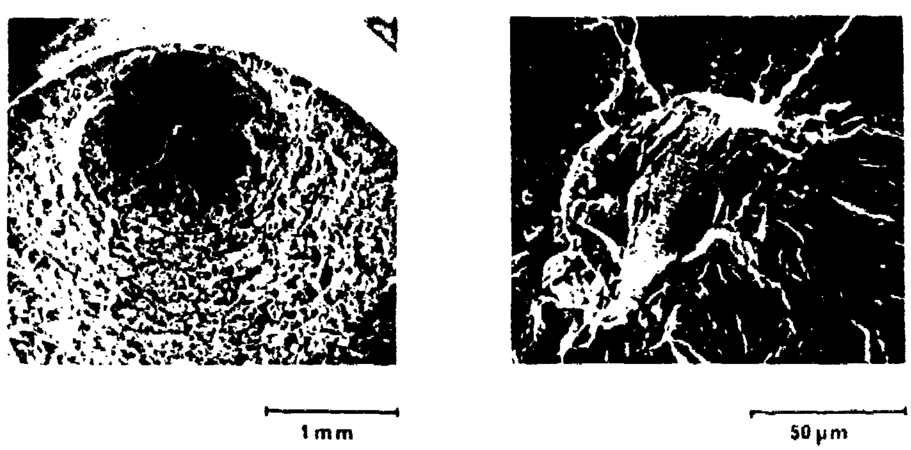


Fig.12 - Defect Fatigue Origin in Astroloy

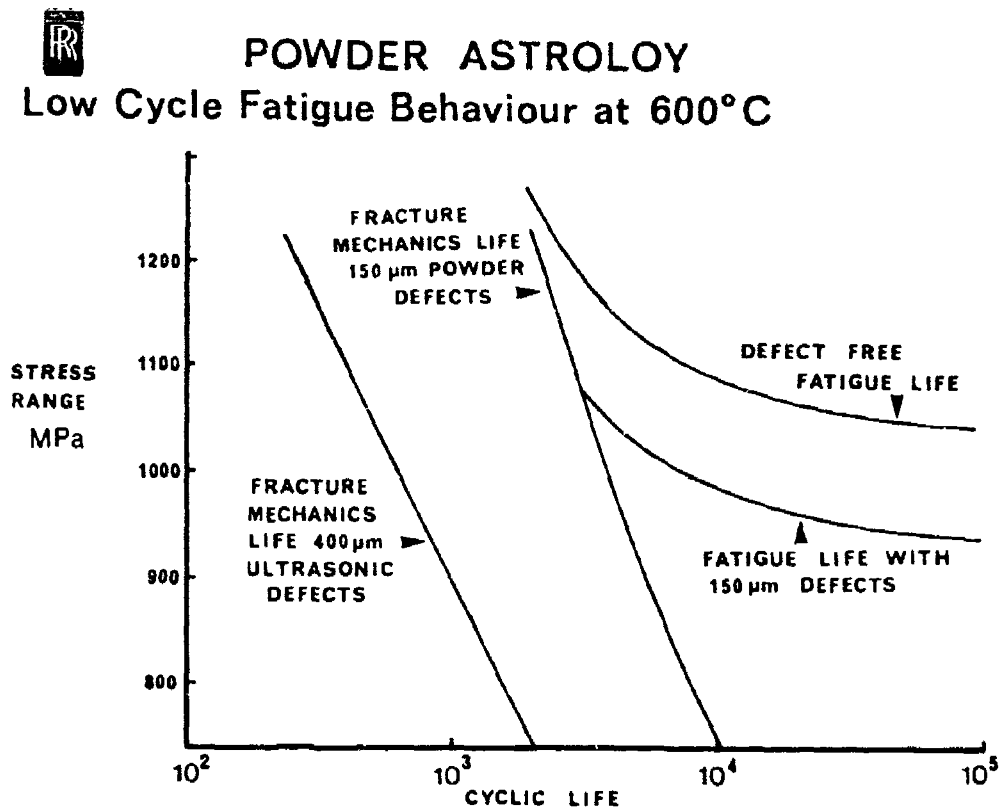


Fig.13 - Cyclic Behaviour of a Powder Superalloy

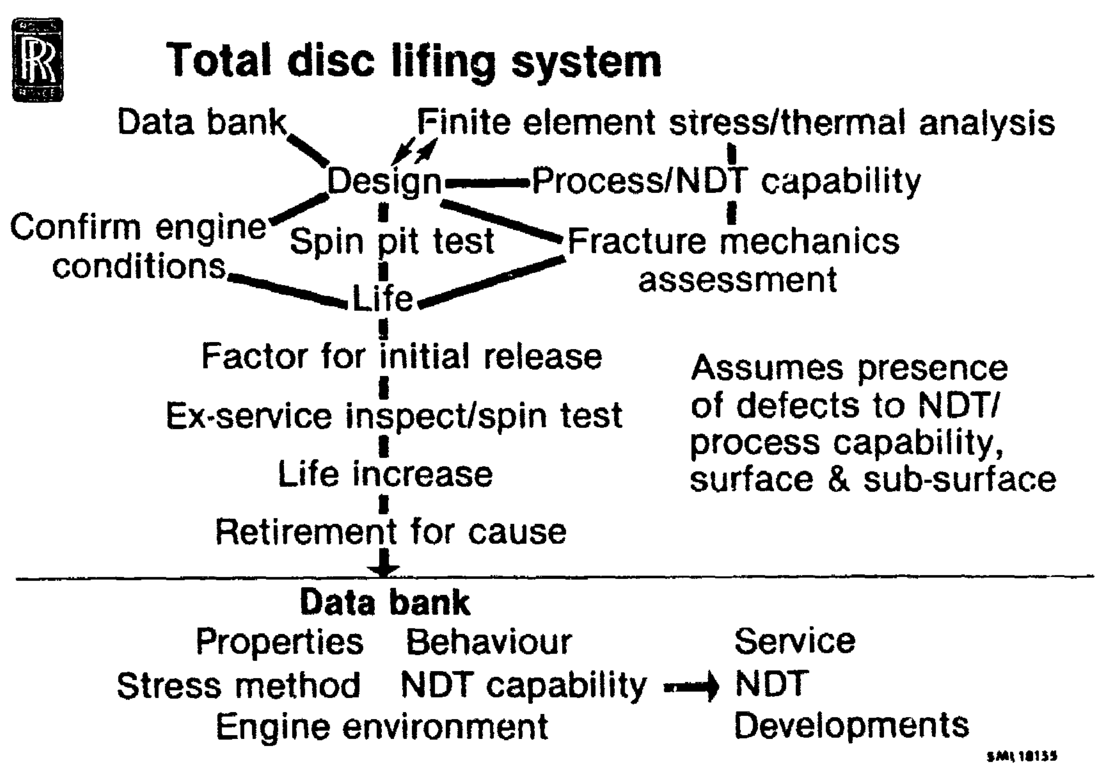


Fig.14 - Total Disc Lifing System

RECORDER'S REPORT – SESSION I

by

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Wg Cdr Hedgecock's paper emphasised the value of recording actual engine useage by means of simple low cycle fatigue counters or more comprehensive monitoring systems. The discussion largely concerned points of detail about such systems.

Replies to various questions can be summarised as follows. The LCF counter in its simplest form takes four shaft speeds as input, possibly two speeds from each of two engines. Speeds are converted to stresses and stress excursions during operation are summed to give the equivalent number of zero/maximum stress cycles which dictate disc life. There is no analogy with the "reference stress" concept used in creep life predictions. Many more inputs (e.g. temperatures) are required for the more elaborate systems which would take into account creep life, thermal fatigue life, etc., and a "hot end monitor" is in process of being developed.

Professor Hoeppner raised the question of the criterion for disc lifing; whether it is "life to first crack" and whether surface and internal defects are treated similarly. It was pointed out that sophisticated Retirement for Cause inspection procedures are not used for RAF engine discs after they have entered service; lives are set by the manufacturers. Formerly lives were prescribed on the basis of "life to first crack" but the current Rolls Royce procedure is based on the more definable value of "⅔ life to burst" obtained from spin-pit tests.

Of the five topics highlighted in Dr Snide's presentation it was one dealing with the Allison electrophoretic coating process that attracted discussion.

In reply to Ir. Mom it was said that the electrophoretic process could be used for applying internal coatings provided that the blade cooling passages were not too small. The major advantage of the process, however, is that it gives a very uniform coating. The thickness after diffusion is typically 0.003 in. This led Ir. Stroobach to raise the important question of the measurement of residual coating thickness after engine service. There appears to be no adequate non-destructive test and the only acceptable method involves cut-up for metallurgical examination.

Several participants (Ir. Mom, Dr Worth, M. Mazars) asked for more details of the coatings and for comparisons with other coating systems. The coating for nickel-base materials, AEP 32, is quoted in the paper as Al-Cr-Mn and is a diffusion-type coating similar to plain aluminide coatings. It is deposited in a propanol + nitromethane bath; details of the bath composition could be made available if required. Test results are given in the paper showing some improvement of AEP 32 over standard aluminide coatings but these results were obtained from test pins and not from engine experience.

Mr Plumb's description of the repair methods used for helicopter engines and naval propulsion gas turbines led Mr Deutsch to ask about the causes of deterioration, particularly the importance of the higher aromatic content of present fuels, a possible higher sulphur content and effects due to erosion by incandescent carbon particles.

In reply it was stated that there are as yet no problems with aero-quality fuels used for helicopters. The diesel fuel used for naval propulsion is rather variable depending on its source, and both higher sulphur and higher vanadium contents are causing some concern. Tests are being done in UK Government Establishments to determine the effects of simulated lower-quality fuels. Carbon erosion, not necessarily from incandescent particles, can be very serious and can only be countered by improving combustor design to reduce carbon formation.

The presentation given by Ir Stroobach drew attention to managing operational and maintenance procedures in order to minimise costs. The main factor is fuel cost; materials costs are small in comparison.

Professor Hoeppner asked for more background information on the engine de-rating schedules that are used. In reply it was said that engines are de-rated as much as possible, the maximum at present being 18%. De-rating at take-off has the largest impact on engine life. Flight crews are unwilling to reduce the take-off thrust below the thrust used during climb and this consideration limits the amount of de-rating that can be used.

The USA programme on Retirement for Cause was described by Mr Harris. A considerable amount of development work, particularly connected with non-destructive evaluation (NDE), is required in order to meet the target date for implementation of the system in 1985. The programme was reported to be on schedule.

Ir. Mom drew attention to a recent proposal that first stage fan discs in the F100 engine should be given a proof test and eddy current inspection after a given number of cycles (perhaps 1800) and then returned to service. This inspect/return sequence, however, would only be permitted for a small number of times (perhaps twice) and then retirement would be mandatory. In contrast, the Retirement for Cause procedure envisaged an indefinite number of inspections and returns-to-service. Did this imply a lack of confidence in the present procedures? In reply it was pointed out that 1985 technology will be an appreciable advance on that of today and it is necessary to proceed cautiously. Further, although proof testing and inspection may confirm that a disc contains no defect likely to cause burst, there are other life-limiting factors such as pitting corrosion or fretting at dovetails which must be taken into account. It is not intended to change any wear or dimensional limits in the Retirement for Cause procedure.

Mr Jeal asked about the basis for selecting the defect size at which a disc would be rejected. The answer was that two approaches would be used, one being the normal fracture mechanics calculation to predict a defect size which would not lead to failure during the next inspection interval, and the other being dependent on the outcome of current work on NDE. Regarding the latter, a probabilistic simulator is being developed which will accept information on detectability and reliability of available NDE techniques and will then be used to select a limiting defect size taking into account appropriate risk and cost criteria.

Professor Hoeppner returned to the question of the criterion for crack initiation; whether this would be taken as "first crack" and whether the same criterion could be used for internally initiated cracks. He also wished to know how the current NDE capability related to the fatigue lifing methodology. In response it was emphasised that the characteristics of the material being used had to be known in very great detail. A large number of discs in IN100 have been cut up for examination and there is now detailed knowledge available concerning the statistics of the sizes and numbers of defects. The Monte Carlo (probabilistic) simulation can thus be used to predict the likely behaviour. For information about internal defects several different NDE methods are used. One of these (eddy current) should reveal just-subsurface defects within about 0.2 in. of the surface. Experience with the inspection of earlier engine discs is not a good basis for developing Retirement for Cause NDE methods since previous engines incorporated less sophisticated designs and ran at lower stresses under less arduous conditions. Detailed inspection of new discs, on the other hand, would be useful; they would be "known" as a result of periodic safety inspections done before the critical inspection on which the decision whether to retire or not is based.

In reply to a further question from Professor Hoeppner it was said that the expected advantages of a Retirement for Cause system would be realised in terms of reduced costs by extending component lives. An alternative advantage could be to increase safety but the present policy is simply not to increase the present risk.

Mr Jeal outlined the more general philosophy being developed in the UK on component lifing and emphasised the considerable amount of detailed information, obtained under realistic conditions, that is required. It would provide a basis for On Condition Lifing where this could be shown to be economically attractive.

Professor Wanhill wanted to know the differences between Retirement for Cause and On Condition Lifing and was told that they are essentially the same.

The more general approach, however, is as much concerned with the first prediction of initial life as with the possibility of life extension by periodic inspection. Both demand a detailed knowledge of the occurrence of defects and their behaviour on a probabilistic basis.

A TITANIUM SILICON COATING FOR GAS TURBINE BLADES

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1. INTRODUCTION

High temperature corrosion in industrial and marine gas turbines is the result of the increased inlet gas temperature of the turbine, which improves the economy, and the use of cheaper fuels, which contain more impurities.
In order to reduce high temperature corrosion, a Ti-Si coating was developed on the laboratory scale by [illegible] [1]. With this coating on IN 738 a number of corrosion tests have been performed [2]. For further development of the coating, and for better understanding the behaviour and degradation of the coating, insight into the corrosion mechanisms is required. A good review is given by Stringer [3]. On the basis of the laboratory development by TNO, the Ti-Si coating has been further developed to produce an industrial coating, Elcoat 360, for use in hot components in gas turbines made of different nickelbase superalloys. Furthermore, other corrosion and oxidation tests have been performed, as a result of which the behaviour and the degradation of the coating is now better understood. In addition tests with Elcoat 360 in gas turbines are still running.

2. COATING PRODUCTION AND COMPOSITION

The first step in the coating process is to make a thin titanium layer on the surface. One of the best methods for this is ion plating [4]. In this process the surface of the components is first cleaned by a glow discharge under low argon pressure. The titanium is then evaporated while the high bias voltage remains. This gives a pure titanium layer of 5 to 15 microns thickness with very good adherence to the base material (see fig. 1a). In the second step, the titanium is diffused into the base material by a heat treatment in vacuum.
As a result of this process, a surface layer of Ni_3Ti is formed with a thickness of about 30 microns (see fig. 1b). In the last step the parts are siliconized by a pack process [5]. In this process the parts are packed in boxes in a mixture of silicon powder, activators such as chlorides and fluorides and inert aluminium oxide as a diluent. These packed coat boxes are heated for 1.5 hours at about 900°C. During this process silicon is transported from the pack to the surface and reacts with the Ni_3Ti layer at the surface. A very complex layer is formed in which G-phase, τ_3-phase and Ni_5Si_3 is formed (see fig. 1c). Fig. 2 shows the change in concentration of the main elements in the coating. Contrary to most of the other coating processes, the heat treatment of the base material is not carried out after coating formation, but is combined with the coating production. The diffusion treatment for the titanium is combined with the solution treatment of the base material. After the siliconizing at 900°C the base material is aged. The deposition of silicon at 900°C causes the gamma prime to tend to overage a little. However, for most of the nickelbase materials no difference in hardness is found compared with the optimally heat treated material.
The reason why the heat treatment cannot be performed after siliconizing is that the coating will melt at temperatures of about [illegible]°C and at temperatures over [illegible]°C the diffusion of silicon into the base material will be noticeable.

3. BEHAVIOUR AND DEGRADATION OF THE COATING

In order to understand the coating behaviour it is important to know the corrosion mechanisms, which can occur in a gas turbine.

3.1. Corrosion mechanisms

The main processes which take place in a gas turbine are oxidation and sulphidation. Dependent on the temperature, three regions can be distinguished each with a different corrosion mechanism (see fig. 3). At relatively low temperatures, about [illegible]°C, acid fluxing sulphidation can occur, caused by a high SO_3 content in the combustion gas. This type of corrosion is characterized by a [illegible] thick sulphidation layer over the whole surface. At temperatures of about [illegible] can occur. A molten layer of sodium sulphate then covers the surface and will [illegible] of the protective oxide scales. There will be a lot of internal sulphidation. At higher temperatures, pure oxidation becomes of more importance. The sodium sulphate will then evaporate.

3.2 Low temperature sulphidation

This type of [illegible] commercially used coatings, [illegible] [illegible]

3.3 [illegible]

[illegible] found at the interface of the oxide and the coating. The [illegible] still contain a certain amount of silicon, but the real protection is given by a thin chromium oxide interlayer. The aluminium oxide also forms an interlayer, while the titanium oxide does not take part in any protection; it only forms discrete particles in the outer part of the oxide scale. These results agree very well with those measured by N. Swindells [illegible]).

3.4 Oxidation

Although the oxidation resistance of the coating is good, less than 5 μm attack during 500 hours oxidation in air at 1000°C, it is not recommendable to use this Elcoat 360 at temperatures over 950°C.
Diffusion of the silicon into the base material and diffusion of different elements from the base material into the coating will occur during exposure at such high temperatures.

3.5 Phase-formation in the coating

The phase in the coating directly after coating production is not fully stabilized. For good stabilization, a heat treatment of about [illegible] hours at temperatures of about 800°C is necessary. In practice this stabilization occurs during normal turbine operation. The concentration distribution of the elements of Elcoat 360 on In 738 directly after coating production is given in Fig. 2. After 300 hours at 850°C the concentrations are changed (see fig. 7). The coating is then stabilized and no further structural changes take place over a longer time.

4. CONCLUSION

It has been shown that the titanium silicon coating Elcoat 360 offers good corrosion resistance against low temperature sulphidation and hot corrosion.

5. ACKNOWLEDGEMENT

The author wishes to thank N. Swindells of the University of Liverpool for his efforts in the X-Ray micro-analyzer work.

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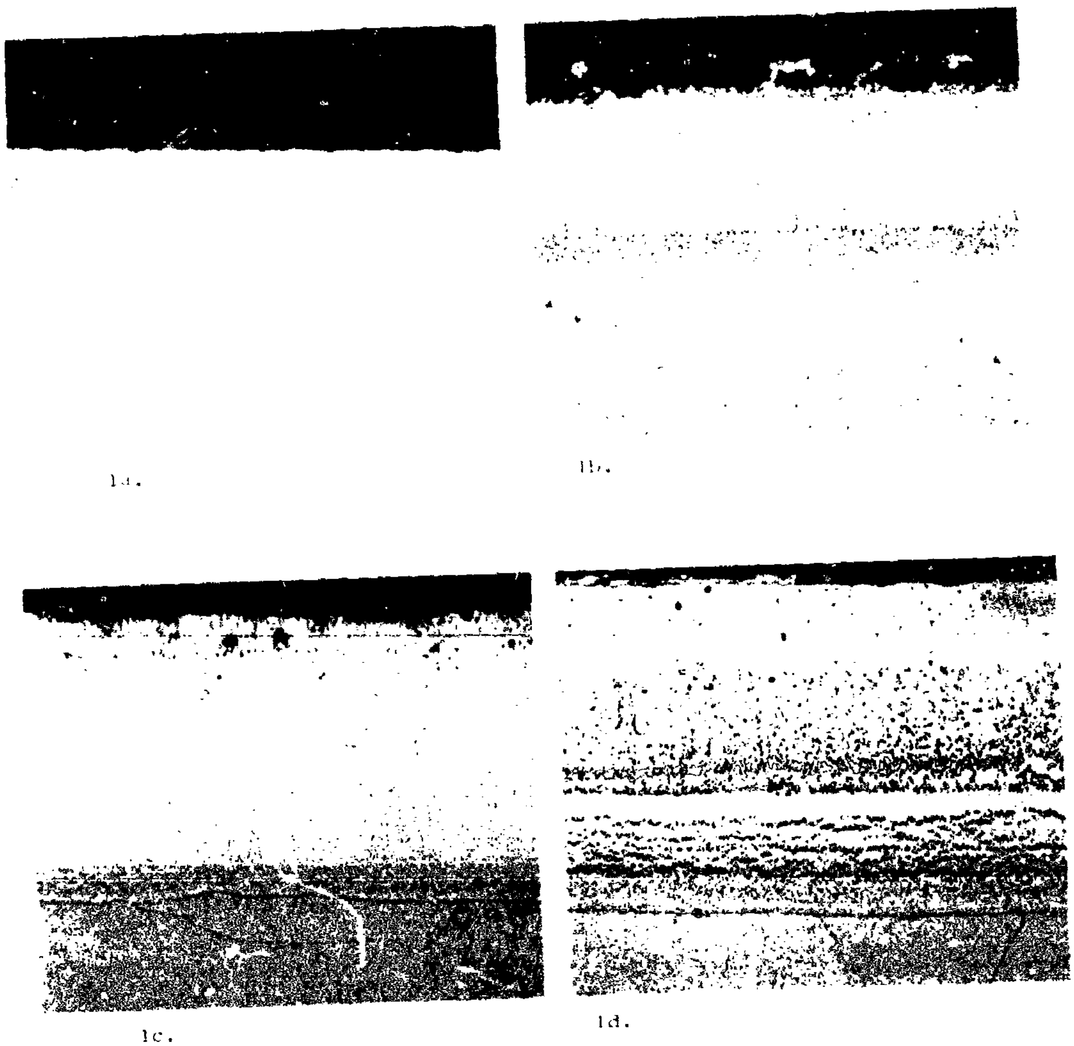


Fig. 1 Different production stages of Elcoat 360 on U 520

a. Ti-plated

b. Ti-diffused

c. after siliconizing

d. after aging

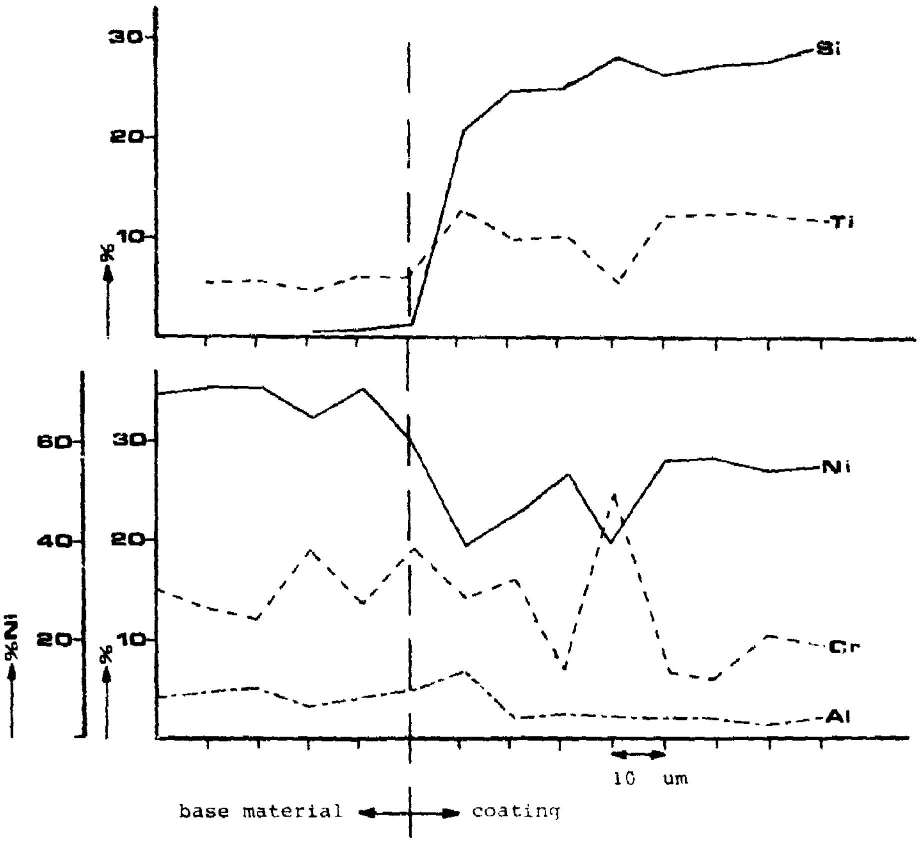


Fig. 2 Distribution of the elements over the coating Elcoat 360 on In 738 LC base material.

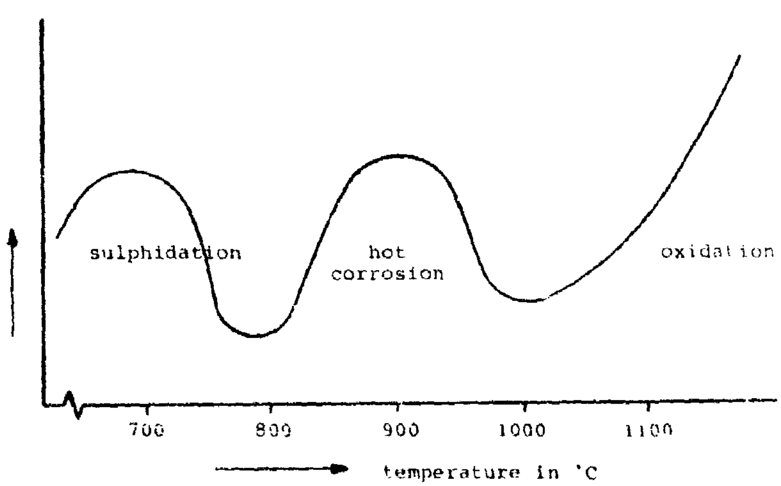


Fig. 3 Possible corrosion mechanisms which can occur in a gas turbine as a function of temperature.

V = 710x

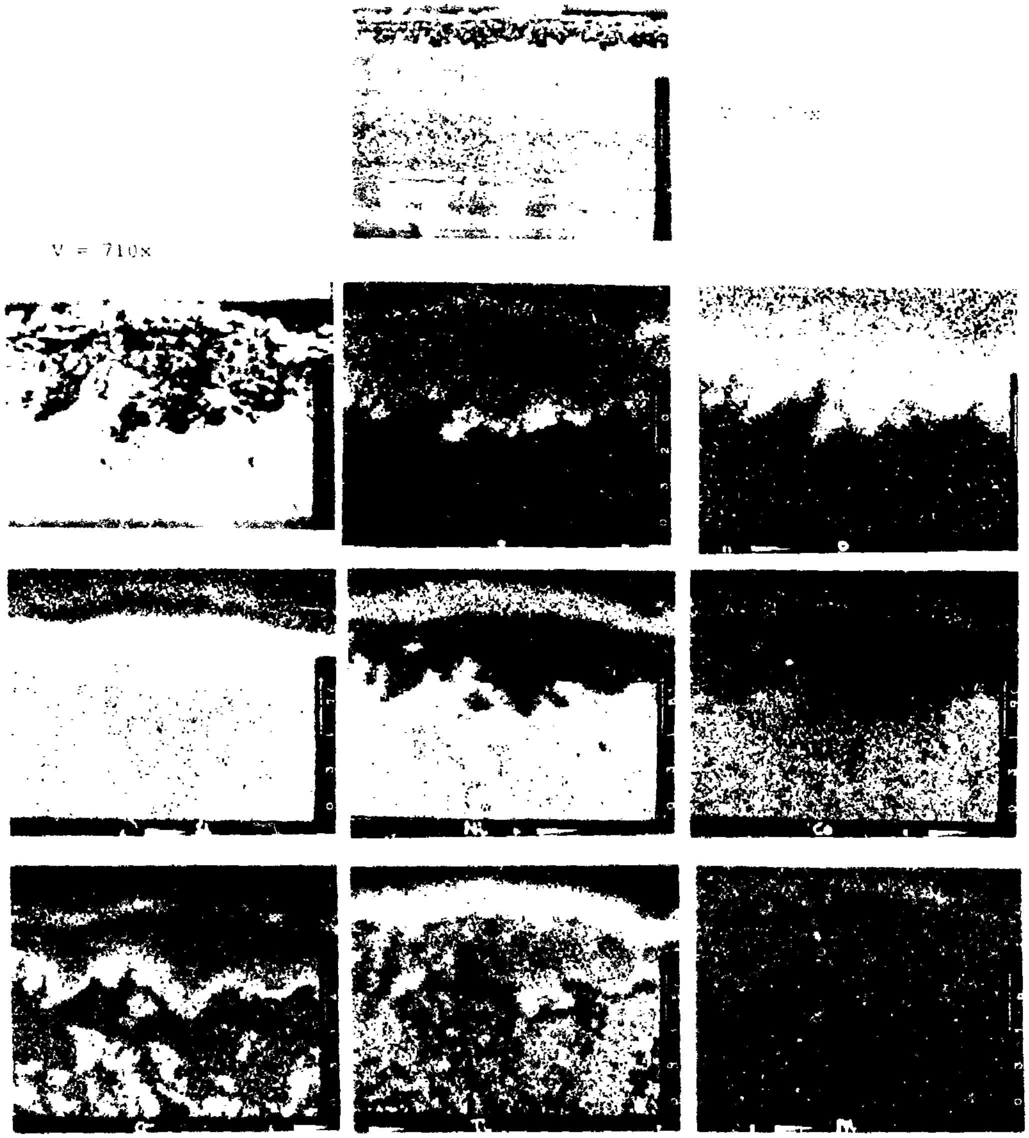
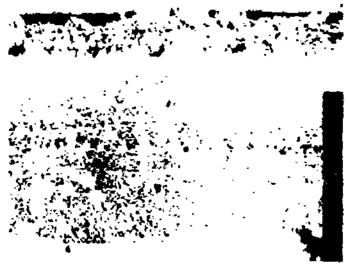


Fig. 4. X-Ray micro analysis of Elcoat 360 on B1900 tested for 1165 hours at 732°C in a burner rig with 2.6% SO_2 and 20 ppm sea salt.



V = 190x

V = 600x

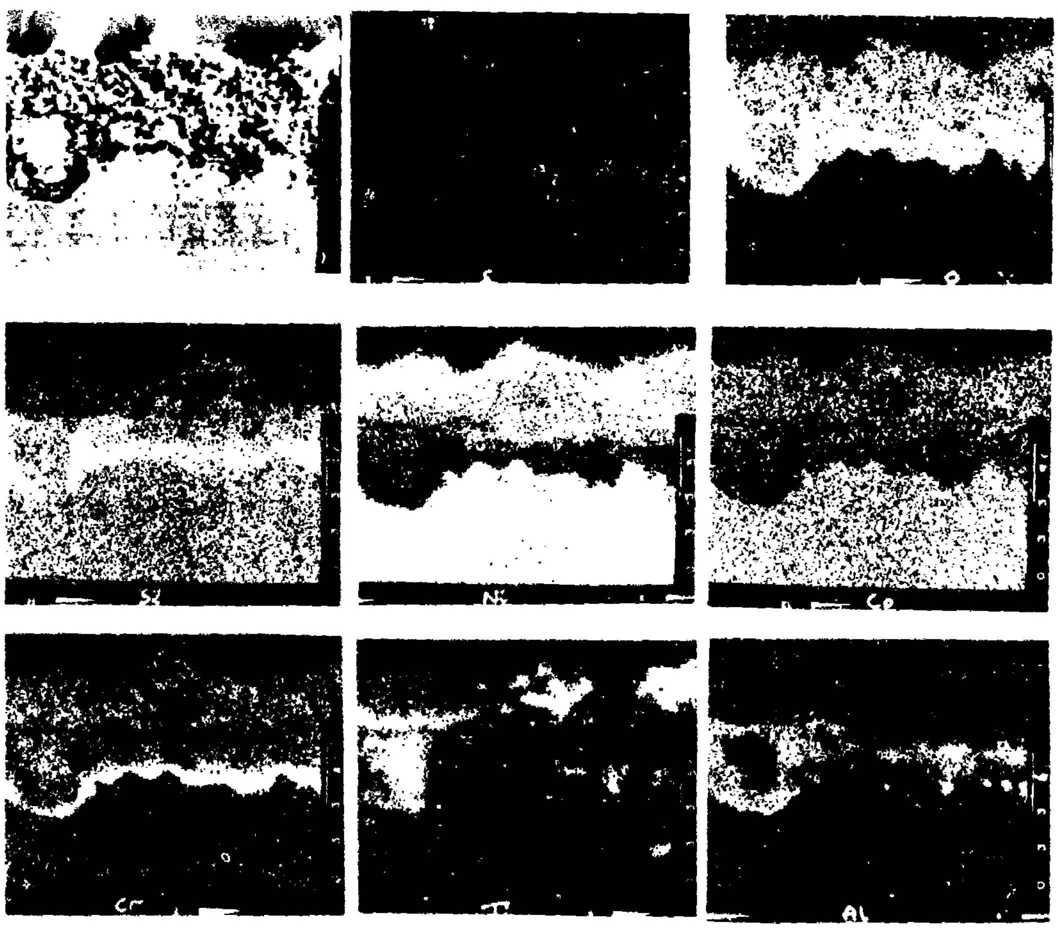


Fig. 5. X-Ray micro analysis of Elcoat 360 on B1900 tested for 1098 hours at 900°C in a burner rig with 1.3% S, and 20 ppm sea salt.

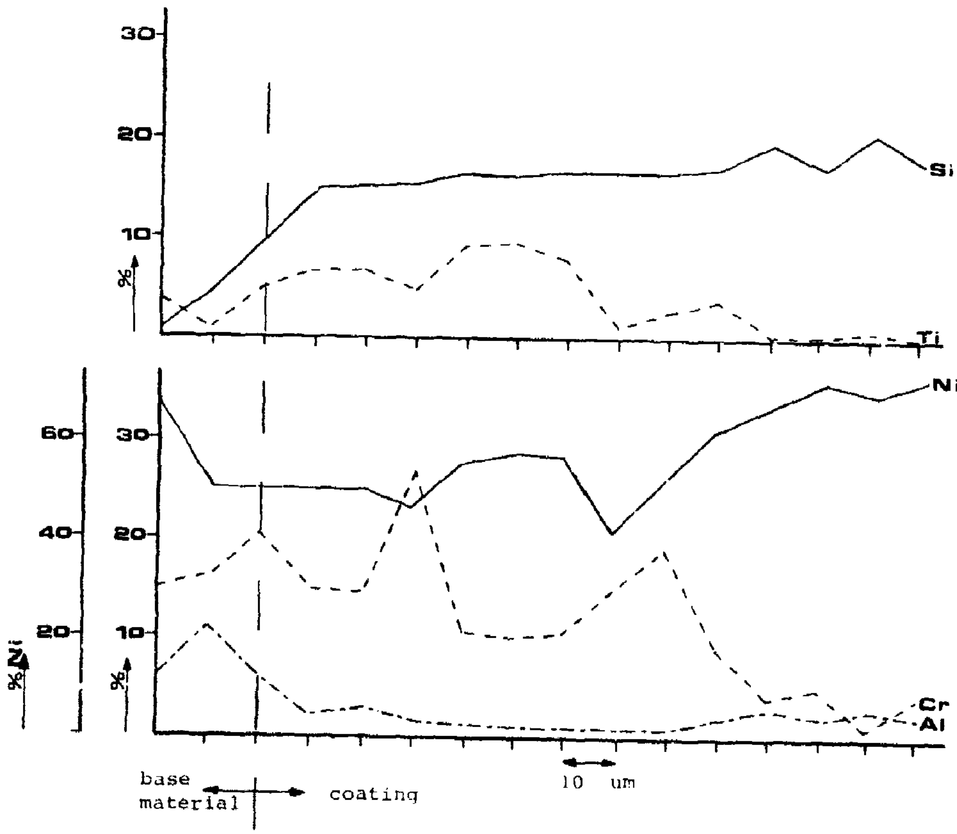


Fig. 6 Distribution of the elements over Elcoat 360 on In 738 LC after 300 hours 850°C in air.

INFLUENCE DES TRAITEMENTS DE PROTECTION SUR LES PROPRIETES MECANIQUES DES PIECES EN SUPERALLIAGE

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1. INTRODUCTION

Les conditions d'environnement des pièces constituant les parties chaudes des turbomachines aéronautiques imposent le plus souvent la réalisation de revêtements protecteurs. Ceux-ci sont d'autant plus nécessaires que la composition des superalliages utilisés et le traitement thermique associé sont choisis pour obtenir les meilleures propriétés mécaniques à chaud. Ce choix de composition ne correspond pas, en général, à une bonne tenue à l'oxydation et à la corrosion ; quant au traitement thermique, il n'est généralement pas compatible avec le cycle thermique de réalisation des revêtements protecteurs et cet aspect est rarement pris en compte par les élaborateurs et même par les bureaux d'études. Pourtant, cycle thermique de protection, caractéristiques mécaniques des matériaux constituant les revêtements, inter-diffusion de ceux-ci avec le superalliage, peuvent avoir une influence défavorable sur les propriétés mécaniques des matériaux protégés : comportement en fluage, fatigue, fatigue thermique ...

La durée de vie des pièces n'est pas seulement déterminée par les propriétés mécaniques du superalliage protégé mais également par la tenue du revêtement à l'oxydation et à la corrosion : renouveler la protection peut poser divers problèmes et, dans certains cas, constituer un nouveau facteur de réduction de durée de vie.

Ces divers aspects seront décrits plus en détail et illustrés par des résultats concernant des revêtements d'aliminiures appliqués aux superalliages IN 100 et IN 738LC. Avant d'exposer les résultats obtenus sur les alliages revêtus, nous avons voulu mettre en évidence l'influence d'un certain nombre de facteurs sur les propriétés mécaniques des superalliages, facteurs inhérents à l'élaboration de la protection.

2. EFFET DU CYCLE THERMIQUE ASSOCIE A L'OPERATION DE PROTECTION

Les opérations de protection, qui comportent le plus souvent au moins un maintien plus ou moins prolongé à température moyenne ou élevée, ont généralement un effet sur la microstructure des superalliages, en particulier sur la précipitation de la phase γ' [Ni_3 (Al,Ti)] des alliages base nickel ou sur la nature des carbures intergranulaires. Peuvent intervenir la durée et la température du palier de traitement, et également, dans certains cas, la vitesse de refroidissement qui reste tributaire de la mise en oeuvre du procédé de protection.

Par conséquent, le traitement de protection peut affecter les propriétés mécaniques de l'alliage non seulement dans la zone superficielle mais également dans tout le volume. Ceci est facilement mis en évidence en soumettant des ébauches d'éprouvettes aux traitements de protection, et en éliminant ensuite par usinage la zone superficielle affectée chimiquement par le traitement.

Dans le cas des revêtements d'alliage de type M-Cr-Al-Y réalisés par projection plasma sous pression réduite, par évaporation thermique ou par des techniques dérivées (ion-plating par exemple), un préchauffage des pièces à température modérée, de l'ordre de 800 °C, est nécessaire pour améliorer l'adhérence du dépôt ; le dépôt est ensuite obtenu dans le même domaine de température en quelques minutes. Dans la mesure où un tel cycle thermique est effectué à température inférieure à celle du traitement thermique de précipitation de la phase γ' (alliages base nickel), il n'a pas d'influence importante sur les propriétés mécaniques du superalliage. Les traitements de post-diffusion à température élevée, destinés à améliorer l'accrochage du dépôt doivent également rester compatibles avec le traitement thermique de l'alliage. On peut, dans certains cas, les faire coïncider avec une mise en solution de la phase γ' de l'alliage (alliages base nickel).

Dans le cas des procédés thermochimiques d'aluminisation, des maintiens de plusieurs heures à température moyenne ou élevée sont toujours nécessaires à l'élaboration de la protection car la cinétique de croissance du revêtement est contrôlée par la diffusion à l'état solide.

- Les traitements d'aluminisation haute activité des alliages base nickel sont effectués assez rapidement à température modérée et conduisent [illegible] à la formation [illegible] de composé [illegible], dont la croissance est contrôlée par la diffusion de l'aluminium [illegible] ; dans une étape suivante de traitement thermique, ce dernier est transformé par [illegible]

et moins fragile (par exemple en 5 h à 1 080°C pour obtenir 60 µm de Ni_2Al_3 sur IN 100). Il est souvent possible de faire coïncider ce dernier traitement avec le traitement de mise en solution de la phase γ' dans les alliages base nickel.

- Les revêtements obtenus par aluminisation basse activité, parfois plus performants du point de vue résistance à la corrosion, ont une cinétique de croissance plus lente, contrôlée par la diffusion du nickel à travers la phase β-NiAl riche en nickel (60 µm de NiAl sur IN 100 en 16 h à 1 050°C ou 60 h à 950°C). L'incidence de tels traitements sur la morphologie des précipités de phase γ' dans les alliages à base nickel, et par conséquent, sur leurs propriétés mécaniques, est généralement importante. De plus, la mise en oeuvre de refroidissement rapide à l'issue des traitements effectués en "pack" est difficile.

- Les traitements de chromisation, qu'il peut être intéressant de réaliser avant aluminisation des superalliages, sont toujours effectués à température supérieure à 1 000°C (30 h à 1 050°C ou 5 h à 1 150°C) et posent les mêmes problèmes de refroidissement que précédemment s'ils sont effectués en "pack".

A titre d'illustration, l'influence de cycles thermiques d'aluminisation sur les caractéristiques de fluage à 850°C de l'IN 100 et de l'IN 738LC, non protégés, est présentée sur les tableaux I et II [1]. On note que cette influence peut varier selon la durée des essais : des essais longs conduisent parfois à une amélioration des durées de vie des éprouvettes traitées, à l'inverse des essais courts (tableau I, lignes 1 et 2). Il est probable que, dans le cas de l'IN 100, la microstructure obtenue après traitement à 1 050°C n'est pas stable et est susceptible d'évoluer lors des maintiens à 850°C en fluage lent, vers une précipitation de γ' plus favorable à la tenue au fluage. Pour évaluer convenablement l'incidence des cycles thermiques de protection sur les propriétés mécaniques des superalliages, il apparaît donc nécessaire de réaliser des essais de fluage de durée suffisamment longue et de ne pas se limiter à des essais de courte durée.

3. CONSOMMATION DE MATIERE LORS DE LA FORMATION DU REVETEMENT

Lors de la formation des revêtements par aluminisation, le superalliage est transformé, par diffusion, sur une certaine profondeur qui dépend de la nature du traitement et de la composition du substrat. Pour les pièces aluminisées, la section utile de superalliage est donc plus faible que la section avant aluminisation. Ceci peut avoir un effet non négligeable sur la tenue en fluage des pièces à parois minces : nous avons en effet montré que les revêtements d'aluminiures ne contribuent pas à supporter la charge de fluage (voir plus loin § 6). Par exemple, la formation d'un revêtement d'aluminiure de 60 µm d'épaisseur entraîne, sur IN 100, une consommation d'alliage d'environ 40 µm ; ainsi, pour une paroi mince d'épaisseur initiale 1 mm, l'épaisseur d'IN 100 restante après aluminisation de chaque face est de 0,92 mm. L'augmentation de contrainte correspondante (X 1,087) à charge constante correspond à une durée de vie réduite d'un facteur voisin de 2 (500 h au lieu de 1 000 h à 850°C). Dans le cas d'une paroi d'épaisseur initiale 0,5 mm, la réduction de durée de vie serait d'un facteur 3 environ, à charge constante également.

Dans le cas de l'élaboration de revêtements par dépôt d'alliage (P.V.D, projection plasma ...), celle-ci n'implique pas une consommation de superalliage mais entraîne seulement une interdiffusion substrat-revêtement limitée. Toutefois, le revêtement d'alliage constitue alors une surcharge (masse supplémentaire soumise à l'accélération centrifuge dans le cas des aubes mobiles) supportée pour l'essentiel par le substrat de superalliage, ceci dans la mesure où l'on admet, comme pour les revêtements d'aluminiures, que le revêtement ne contribue pas à supporter la charge de fluage.

Finalement, pour les deux types de revêtements, les rapports de la section totale des pièces déterminant la charge due à l'accélération centrifuge à la section de superalliage supportant effectivement cette charge sont sensiblement comparables.

4. INTERDIFFUSION EN SERVICE ENTRE LE REVETEMENT ET L'ALLIAGE

Lors de maintiens en service à haute température, de nouvelles phases peuvent se développer par interdiffusion entre le revêtement et le substrat. Ces phases peuvent éventuellement jouer un rôle spécifique dans la tenue mécanique des pièces, selon leur morphologie (couche continue, précipités dans le revêtement, dans l'alliage sous-jacent au revêtement), et selon leurs propriétés (ductilité, fragilité). Cette morphologie dépend essentiellement de la nature du substrat et du mode d'élaboration du revêtement (revêtement d'aluminiure obtenu par diffusion, revêtement obtenu par dépôt d'alliage). Dans le premier cas, on obtient généralement à la fois des précipitations inter et intragranulaires et des couches continues de phases intermédiaires, alors que dans le deuxième cas, par un choix approprié de la composition de l'alliage du dépôt, l'interdiffusion peut être limitée à une variation progressive de composition sans apparition de phases "fragiles".

4.1. Interdiffusion entre un revêtement d'aluminiure et le superalliage

Les revêtements d'aluminiure, formés à la fois par apport d'aluminium par voie gazeuse et par transformation de la zone superficielle du superalliage, ne peuvent mettre en solution qu'une faible partie des éléments d'addition de ce dernier, éléments qui précipitent dans la zone interne sous forme de carbures ou phases intermédiaires (σ par exemple) riches en Cr, Mo, W, Co ... et ont tendance à diffuser dans la zone sous jacente du substrat au cours de maintiens en température.

Selon l'alliage et la température de maintien, on note l'apparition de couches monophasées γ et γ' et/ou des précipitations intergranulaires (carbures) ou intragranulaires (σ, carbures ...) sur une profondeur pouvant atteindre 200 µm. Ainsi, dans le cas de l'IN 100 aluminisé, on observe des précipitations intergranulaires de carbures de chrome sur 150 µm sous le revêtement après 300 h de maintien à 850°C (figure 1a), ou sur 250 µm sous le revêtement après 300 h à 1 050°C (figures 1b et 1c), ainsi qu'une précipitation intragranulaire de phase aciculaire σ (sur 20 µm après 300 h à 850°C) ou de carbures de chrome (sur 50 µm après 300 h à 1 050°C).

Dans le cas du fluage, il a été observé [1] [2] que des fissures amorcées dans le revêtement à un stade avancé de fluage peuvent se propager le long des plaquettes σ formées dans l'alliage (figure 2). Néanmoins, il n'a jamais été observé de ruptures à partir de telles fissures, ces ruptures ayant toujours à l'origine des décohésions intergranulaires nombreuses dans l'IN 100 soumis au fluage [1] et ceci même au voisinage immédiat de fissures se propageant sur les plaquettes de phase σ intragranulaires (figure 3). Les précipitations intergranulaires de carbures de chrome dans le substrat au voisinage du revêtement d'aluminiure pourraient, par contre, dans ces conditions, avoir une influence sur le comportement en fluage (figure 4 : observation d'une décohésion à l'interface carbure, grain $\gamma - \gamma'$). Il faut ajouter que dans le cas où se forment des couches monophasées γ et γ', celles-ci ne contribuent pas à supporter la charge de fluage.

Dans le cas des sollicitations de fatigue, la précipitation aciculaire de phase σ pourrait avoir un effet plus déterminant : en effet, il a été montré [3] que, après des recuits de 5 000 h à 850°C, l'INCO 738LC aluminisé par le procédé LDC2 présente une baisse de durée de vie en fatigue vibratoire à 850°C que l'on peut corréler avec un amorçage de fissures dans la zone de précipitation de phase σ.

4.2. Interdiffusion entre un revêtement d'alliage et le superalliage

L'interdiffusion entre le revêtement de type MCrAlY et le substrat peut être très réduite car, d'une part la diffusion est limitée pendant le processus de dépôt, et d'autre part un choix approprié de la nuance de revêtement permet d'obtenir rapidement une bonne stabilité des phases en présence.

Dans certains cas, cependant, l'interaction peut être importante ; par exemple, dans le cas des revêtements de type CoCrAlY déposés sur un superalliage base nickel, la fraction volumique de la phase β riche en alumium diminue progressivement, tandis que se forme une couche intermédiaire de phase γ' Ni_3Al. De telles couches d'interdiffusion ne présentent pas de fragilité particulière, mais ne contribuent pas à supporter la charge de fluage.

5. PROPRIETES MECANIQUES DES PHASES CONSTITUANT LE REVETEMENT

Les revêtements obtenus par aluminisation comprennent généralement une matrice d'aluminiure (β - NiAl) et des précipités ; les revêtements d'alliage sont constitués de phase γ - cfc base Co ou Ni, riche en Cr, généralement prépondérante, et de phase β - NiAl (ou Co Al). Le caractère fragile à basse température des intermétalliques NiAl et CoAl détermine le comportement des revêtements dans les deux cas ; dans les revêtements d'alliage, le caractère fragile est atténué par la plasticité de la phase γ qui constitue, en général, la matrice du revêtement.

La température de transition ductile-fragile des revêtements d'aluminiure a été étudiée, en particulier en traction à l'aide de méthodes d'émission acoustique [4]. Pour la phase NiAl, riche en aluminium, elle est de l'ordre de 650 à 750°C ; au-dessous de 650°C, des fissures apparaissent en traction pour des élongations de 0,5 à 1,5% ; au-delà de 750°C, les revêtements d'aluminiure peuvent être au contraire plus ductiles que le superalliage et peuvent subir des allongements importants sans fissurer (figure 5). Les aluminiures de cobalt apparaissent comme plus fragiles que les aluminiures de nickel : allongement à rupture en traction de l'ordre de 0,25 % au-dessous de 800°C.

Les propriétés mécaniques des revêtements d'alliage dépendent des propriétés individuelles des deux phases constitutives ainsi que de leurs proportions et de leur morphologie. Dans le cas d'alliages binaires biphasés γ/β (Co-CoAl), recristallisés, une interprétation du comportement mécanique à l'ambiante a pu être donnée en utilisant un modèle élastique pour la phase fragile β et élasto-plastique pour la matrice γ ; le comportement change lorsque la phase β devient continue, c'est à dire lorsque sa teneur dépasse 28 % en volume : phase disposée aux joints de grains. De même, on observe que des alliages CoCrAlY à morphologie globulaire, obtenus par évaporation thermique, présentent une variation de température de transition ductile-fragile autour d'une teneur en aluminium de 11% en masse (soit environ 45% de phase en volume), c'est à dire, dans le cas de cette morphologie, selon que la phase β est continue ou non (figure 6 [5]).

6. PROPRIETES MECANIQUES DES PIECES EN SUPERALLIAGE [illegible]

A basse température, les aluminiures [illegible] fragiles, ce qui peut détériorer la tenue à l'impact, à la fatigue thermique et à la fatigue [illegible] ; les problèmes se posent différemment à haute température où les aluminiures sont [illegible].

6.1. [illegible]

[illegible]

Les essais de longue durée [2] (quelques dizaines de milliers d'heures) à 850°C sur INCO 738LC et FSX 414 aluminisés, ayant subi le traitement thermique standard après aluminisation, montrent que la présence du revêtement est sans effet sur la durée de vie -la contrainte étant calculée sur la section de l'éprouvette avant aluminisation-.

Les essais de courte et moyenne durée à 850°C sur IN 100 et INCO 738LC aluminisés [1] montrent que ce n'est pas la présence du revêtement en elle-même qui modifie la durée de vie du fluage, le temps d'allongement à 1% et la vitesse de fluage secondaire (figure 7) - la contrainte étant calculée sur la section résiduelle de superalliage après aluminisation -. En effet, la comparaison d'essais sur éprouvettes nues avant protection, protégées, nues mais ayant subi le cycle thermique de protection, met en évidence que la dégradation des propriétés mécaniques observées dans les deux derniers cas pour les essais courts, doit être imputée au cycle thermique de protection et non au revêtement lui-même (cf. §2).

Ces deux études tendent donc à montrer que la présence du revêtement est sans effet sur la durée de vie en fluage.

Le fait de ne pas calculer la contrainte de la même façon dans les deux cas ne modifie en rien la conclusion précédente car, pour les éprouvettes utilisées -éprouvettes cylindriques de quelques millimètres de diamètre-, les corrections de contrainte sont faibles et leur effet sur la durée de vie s'inscrit sensiblement dans la dispersion des résultats. Il a été toutefois montré par des essais multiples [1] que le revêtement d'aluminiure subit la déformation imposée par la partie centrale en superalliage sans porter une part appréciable de la charge et, de ce fait, il reste préférable de faire la correction de contrainte ; par contre, cette correction est indispensable dans le cas des éprouvettes à parois minces.

L'observation des revêtements d'aluminiures après fluage à 850°C [1] montre que ceux-ci présentent une plasticité importante en fluage à cette température : à 4% de déformation, les revêtements sont encore exempts de fissures ; des déformations encore plus importantes sont possibles localement, en particulier au droit des cavités intergranulaires qui s'ouvrent dans le superalliage (figure 5). Après une déformation globale importante, des fissures apparaissent dans le revêtement au niveau des zones de striction. Pratiquement, de telles fissures ne sont jamais rencontrées sur pièces réelles, puisque celles-ci sont retirées du service après 1 à 2% de déformation ; leur rôle éventuel sur la rupture a cependant été discuté (§ 4), puisque leur propagation a été observée sur quelques dizaines de µm dans le superalliage le long des plaquettes de phase σ de la zone d'interdiffusion et il a été montré que, dans nos conditions d'essais, ces fissures ne semblent pas initier la rupture qui reste intergranulaire dans le superalliage (figure 8). Il parait important de rappeler que, par contre, étant donné la nature de cette rupture, les phases (carbures) développées par interdiffusion revêtement/substrat dans les joints de grains de ce dernier peuvent influencer le comportement en fluage, en particulier dans le cas des pièces à parois minces.

6.2. Fatigue

La résistance à la fatigue est une propriété particulièrement sensible à l'état de surface et à l'environnement, puisque l'amorçage des fissures a souvent lieu à la surface des pièces. Les revêtements peuvent donc modifier de façon notable la tenue de ces dernières si leur résistance chimique à l'environnement permet d'éviter les interactions corrosion/fatigue ou oxydation/fatigue auxquelles peut être sensible le superalliage. Il en est de même s'ils présentent une fragilité particulière dans certains domaines de température ou pour certains types de sollicitations de fatigue.

Le rôle spécifique du revêtement, en fonction de sa nature et de la température, a seulement pu être mis en évidence sur des matériaux tels les superalliages solidifiés unidirectionnellement poly ou monocristallins (exemple DS 200 [6]), les alliages forgés (UDIMET 700 [6]) ou les superalliages composites à fibre de carbure obtenus par solidification unidirectionnelle [7], alliages moins sensibles à la fatigue que les alliages coulés équiaxes. En effet, dans le cas des alliages coulés equiaxes, ce sont des amas de porosités internes qui provoquent généralement l'amorçage de la rupture (figures 9.10.11), ce qui rend les revêtements sans effet sur la durée de vie en fatigue (850°C, grand nombre de cycles, figure 12).

Selon le type de sollicitations de fatigue, il faut distinguer : la fatigue à grand nombre de cycles à la rupture (HCF) et par conséquent faible taux de déformation, la fatigue à fort taux de déformation et par suite faible nombre de cycles à rupture (LCF). Dans tous les cas envisagés ci-après, la déformation du revêtement est imposée par celle du substrat.

Fatigue à faible taux de déformation

La déformation du substrat de superalliage est alors essentiellement élastique et, d'une manière générale, les matériaux constituant le revêtement auront tendance à avoir une bonne tenue en fatigue s'ils ont une limite élastique élevée ou s'ils ont un faible module d'élasticité. Dans ce dernier cas, en effet, pour un taux de déformation imposé, pour deux matériaux dont le module d'élasticité diffère, le rapport des contraintes qu'ils supportent est égal au rapport de leur module d'élasticité respectifs.

Ainsi un revêtement d'alliage Ni-20Cr-10Al-2Hf-0,1C sur NiTaC 3 - 116 A2 (eutectique orienté contenant des fibres de TaC) a une meilleure tenue en fatigue qu'un revêtement d'alliage Ni-20Cr [7] dont la limite élastique est vraisemblablement plus faible, ce qui peut même conduire à ce qu'il soit sollicité dans le domaine plastique. De même, à faible taux de déformation, un revêtement même fragile pourvu qu'il soit sollicité en deçà de son domaine de déformation plastique n'a pas d'influence sur la tenue en fatigue : par exemple, il a été observé [6] que l'aluminisation de l'UDIMET 700 -suivie de polissage électrolytique pour annuler tout effet de concentration de contrainte- non seulement ne détériore pas mais améliore le comportement en fatigue à l'ambiante de l'UDIMET 700 non revêtu et également poli. Inversement, lorsque le revêtement subit une déformation plastique alors que le substrat est encore dans le domaine élastique, on peut s'attendre à un abaissement notable de la durée de vie.

Fatigue à taux de déformation élevé

Dans ce cas, le substrat et le revêtement sont sollicités au-delà de leur limite élastique.

Si le revêtement est fragile (aluminiure, alliages γ-β à basse température), des fissures peuvent y apparaître dès le premier cycle et réduire notablement la durée de vie en fatigue. Ainsi, la durée de vie de UDIMET 700 est réduite de 20% à 760°C lorsqu'il est revêtu d'un aluminiure [6].

Au contraire, si le revêtement est ductile (aluminiure ou alliages γ-β à hautes températures, alliage Ni-20Cr), la tenue en fatigue du superalliage n'est pas dégradée et peut même être améliorée. Par exemple, dans le cas du NiTaC 3 - 116 A2, alliage dans lequel il a été observé [8] que la création volontaire de fissures à la surface réduisait notablement sa durée de vie en fatigue, le revêtement Ni-20Cr-10Al-2Hf-0,1C, relativement fragile, en se fissurant au bout d'un petit nombre de cycles, peut réduire considérablement la durée de vie de l'alliage [7], alors que le revêtement Ni-Cr, ductile, n'aura pas cette influence néfaste.

7. CONCLUSION

Quelques aspects de l'influence des revêtements protecteurs sur les propriétés mécaniques des pièces en superalliages ont été discutés : notamment cycle thermique de protection, interdiffusion entre le revêtement et le superalliage, propriétés physiques et mécaniques des revêtements eux-mêmes.

Certains des effets décrits sont considérablement amplifiés dans le cas de la protection des pièces à parois minces fortement refroidies ; les problèmes d'interdiffusion revêtement/superalliage, d'influence des revêtements sur le comportement en fatigue thermique notamment, sont particulièrement aigus et il n'existe pas actuellement de solution satisfaisante.

D'autres aspects auraient pu être pris en compte dans ce texte ; ainsi la recherche de caractéristiques mécaniques élevées pour les superalliages, soit pour augmenter la température d'emploi des pièces, soit pour augmenter leur potentiel, conduit à des compositions et des traitements thermiques d'alliages bien déterminés ; ces choix précis peuvent conduire à limiter ceux de la nature des revêtements et de la méthode de réalisation, entraînant ainsi une tenue insuffisante de ces revêtements à l'oxydation et à la corrosion et par suite une limitation de la durée de vie des pièces. Leur réparation par renouvellement du revêtement est courante mais pose divers problèmes : difficultés de décapage pour les revêtements d'alliages, diminution de section des pièces lors du décapage des revêtements d'aluminiures correspondant à la partie de superalliage consommé pour la réalisation de ces revêtements et limitant le nombre d'opérations possibles de renouvellement.

Les résultats présentés dans le cas des alliages IN 100 et IN 738LC ne peuvent pas être généralisés à d'autres alliages ; chaque couple revêtement/superalliage doit faire l'objet d'investigations particulières dans des conditions aussi voisines que possible de celles de l'utilisation des pièces.

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Tableau I – Influence des cycles thermiques d'aluminisation IN 100 nu – fluage à 850° C sous air.

Cycle de traitements	ESSAIS COURTS			ESSAIS MOYENNE DUREE		
	σ (MPa)	* $t_{1\%}$ (h)	* $t_{rupture}$ (h)	σ (MPa)	* $t_{1\%}$ (h)	* $t_{rupture}$ (h)
1. Brut de coulée	372,8	116	200	294,3	515	927
2. 16h à 1050°C refroidissement 35°C/mn (aluminisation basse activité sur ébauche)	372,8	65	182	294,3	574	1179
3. idem 2 + 300 cycles de 1h à 850°C	372,8	74	222			
4. idem 2 + 300 cycles de 1h à 1050°C	372,8	71	151			
5. 7,5h à 700°C + 5h à 1080°C puis trempe à l'air (aluminisation haute activité sur ébauche)	372,8	107	294	294,3	622	935

*temps moyens

Tableau II – Influence des cycles thermiques d'aluminisation INCO 738 LC nu – fluage à 850° C sous air.

Cycle de traitements	ESSAIS COURTS		
	σ (MPa)	* $t_{1\%}$ (h)	* $t_{rupture}$ (h)
1. Brut de coulée	294	98	226
2. 16 h à 1050°C refroidissement 35°C/mn (aluminisation basse activité sur ébauche)	294	54	312
3. 2 h à 1120°C refroidissement 35°C par mn + 24h à 845°C trempe à l'air (traitement thermique standard de l'alliage)	294 310	50 (1) 32	
4. idem 3. + 16 h à 1050°C refroidissement 35°C par mn (aluminisation basse activité sur ébauche) + idem 3.	310	30	
5. idem 3. + 16h à 1050°C refroidissement 35°C par mn (aluminisation basse activité sur ébauche) + 24h à 845°C trempe à l'air	310	24	
6. idem 3 + 7,5h à 700°C (aluminisation haute activité sur ébauche) + idem 3	310	41	

* temps moyens (1) valeur interpolée

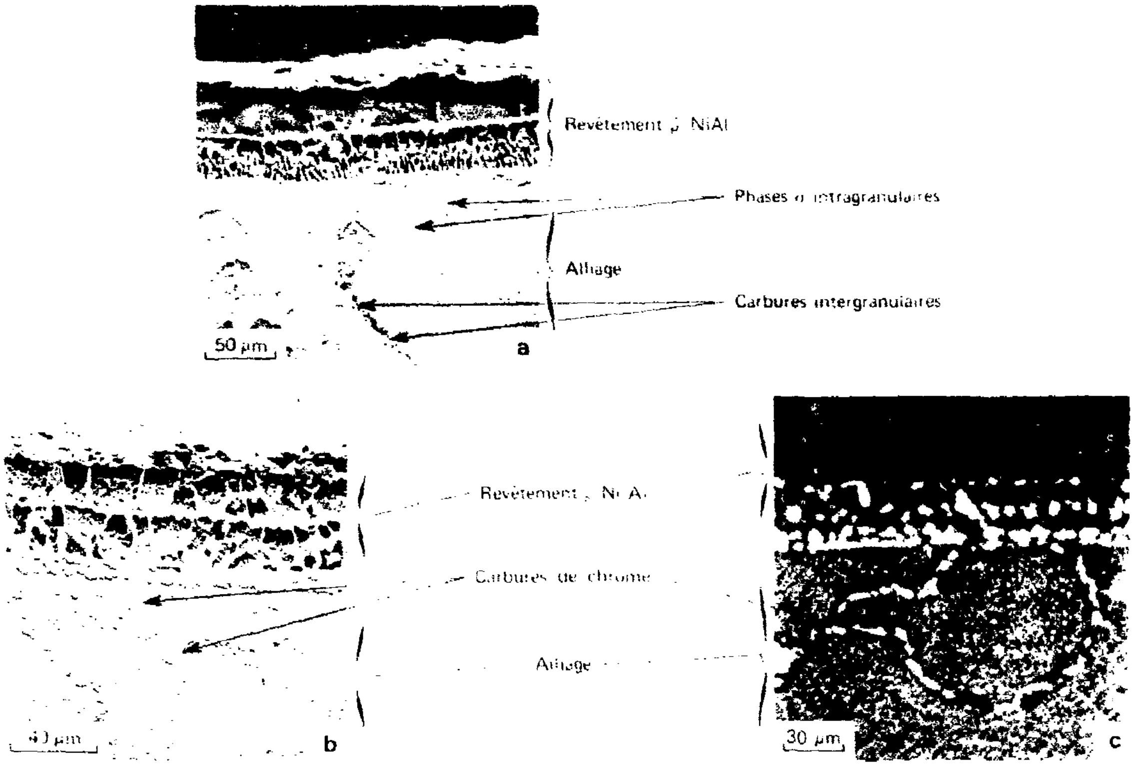

Fig. 1 – Interdiffusion revêtement (aluminisation basse activité) – substrat
a) 300 cycles de 1h à 850° C sous Ar (image électronique)
b) 300 cycles de 1h à 1050° C sous air (image électronique)
c) 300 cycles de 1h à 1050° C sous Ar (image X du chrome).

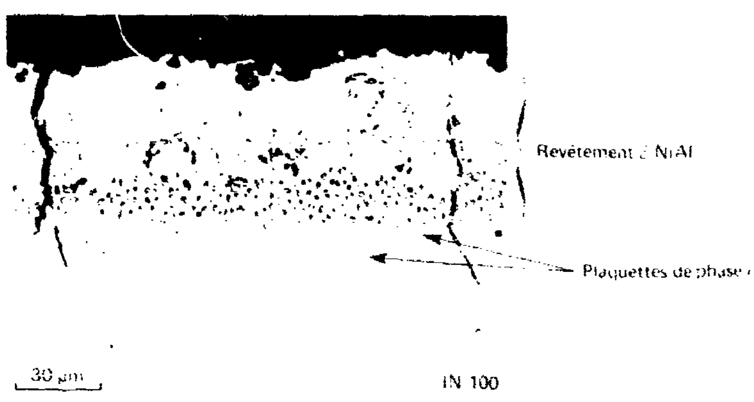

Fig. 2 – IN 100 aluminisé (basse activité) : fluage à 850 C sous air (388 MPa, rupture en 232 h) – phases σ formées par interdiffusion substrat/revêtement pendant le maintien à 850° C en fluage.

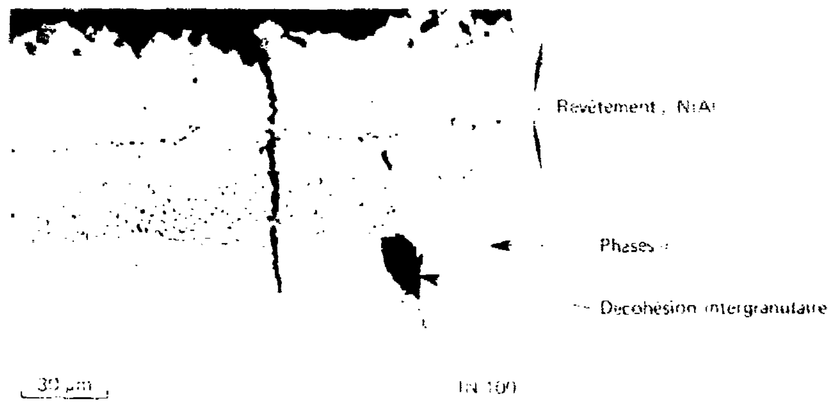

Fig. 3 — IN 100 après aluminisation basse activité, traitement de 300h à 850 C sous air (300 cycles de 1h), fluage à 850° C sous air (388 MPa, rupture en 984 h).

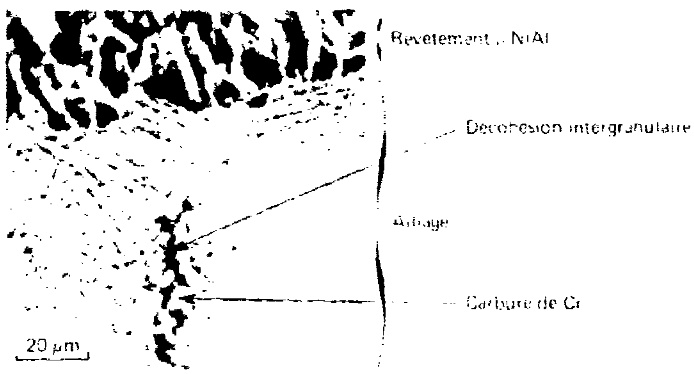

Fig. 4 — IN 100 après aluminisation basse activité, traitement de 300h à 850 C sous argon (300 cycles de 1h), fluage à 850° C sous air (σ 388 MPa, rupture en 235h — image electronique).

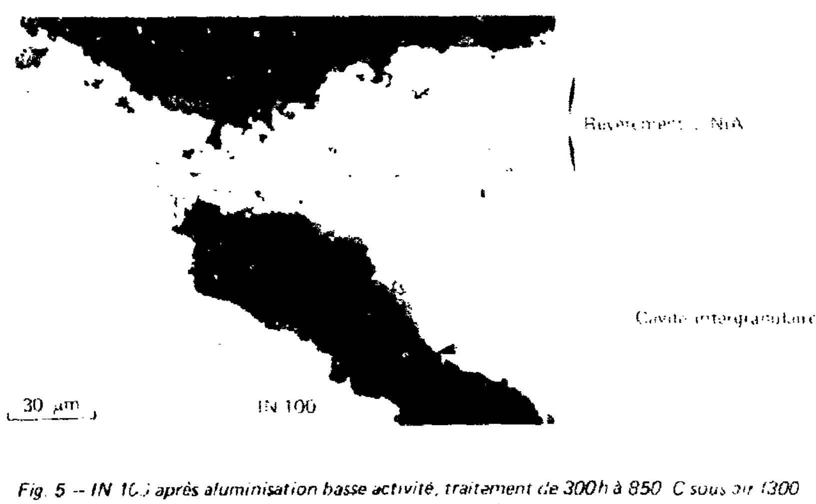

Fig. 5 — IN 100 après aluminisation basse activité, traitement de 300h à 850 C sous air (300 cycles de 1h), fluage à 850 C sous air (σ 388 MPa, rupture en 98 h).

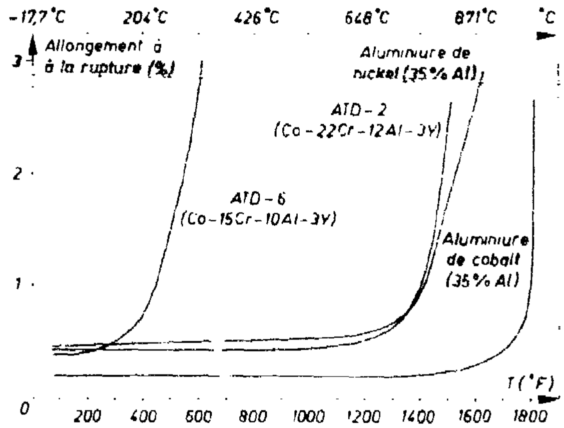

Fig. 6 – Evaluation de la température de transition ductile-fragile d'alliages Co Cr Al Y et d'aluminiures Ni Al et Co Al (d'après [5]).

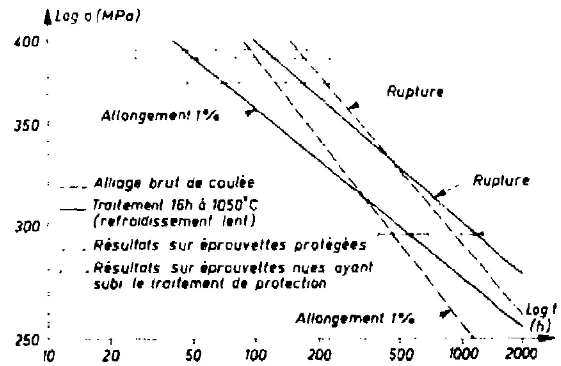

Fig. 7 – Fluage à 850° C sous air : log σ = f (log t) IN 100 brut de coulée et traité 16 h à 1050° C (cycle thermique de protection).

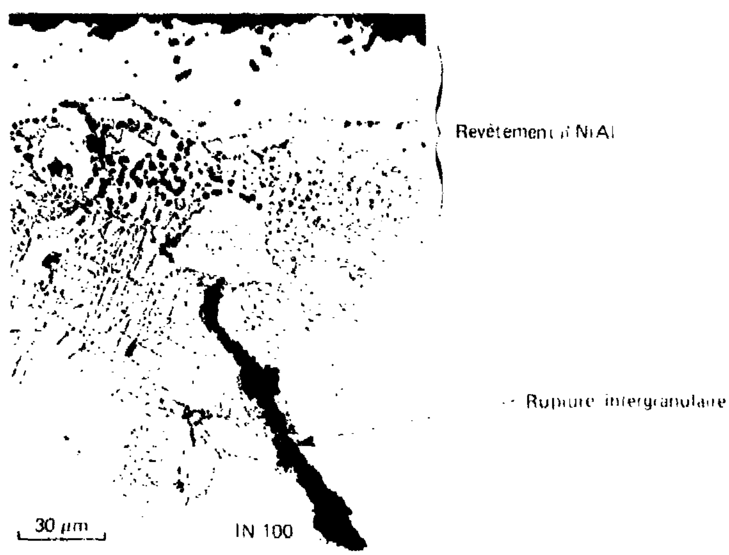

Fig. 8 – IN 100 après aluminisation basse activité, traitement de 300 h à 850° C sous air (300 cycles), fluage à 850° C sous air (388 MPa, rupture en 98 h).

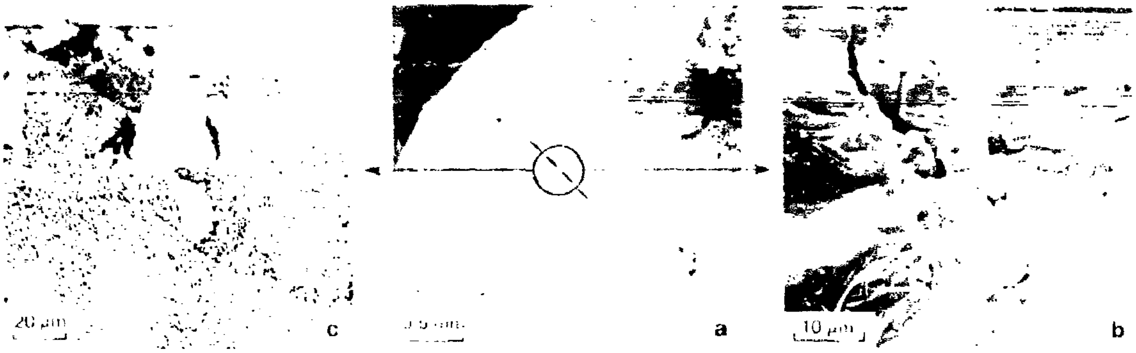

Fig. 9 – IN 100 non revêtu. Fatigue à 850 C. 45 – 450 MPa, rupture en 240 000 cycles.

a) surface de rupture (MEB)
b) amorce de rupture (MEB)
c) vue en coupe de l'amorce
d) autre amorce, à 2 mm du plan médian (coupe longitudinale).

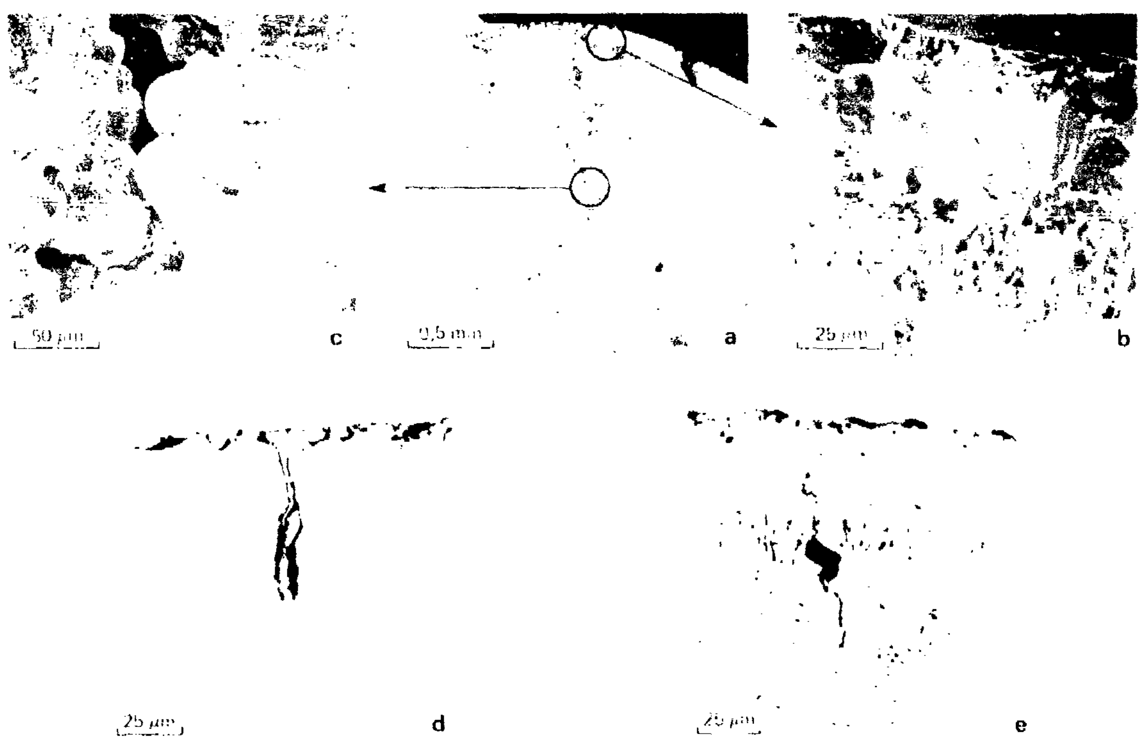

Fig. 10 – IN 100 chromaluminisé. Fatigue à 850° C, 45–450 MPa, rupture en 115 000 cycles.

a) surface de rupture (MEB)
b) surface de rupture du revêtement (MEB)
c) amorce (MEB)
d) et e) coupes longitudinales du revêtement.

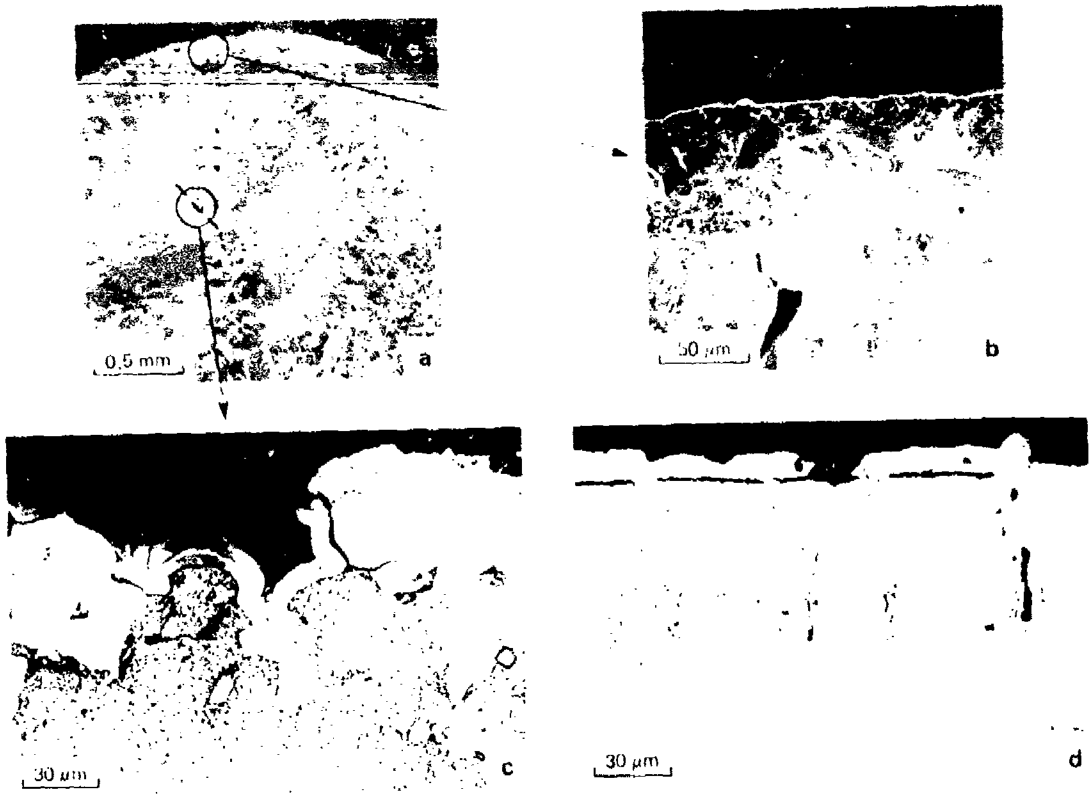

Fig. 11 – IN 100 aluminisé haute activité. Fatigue à 850 C, 45 – 450 MPa, Rupture en 66 000 cycles.

a) surface de rupture (MEB)
b) surface de rupture du revêtement (MEB)
c) coupe de l'amorce
d) coupe longitudinale du revêtement.

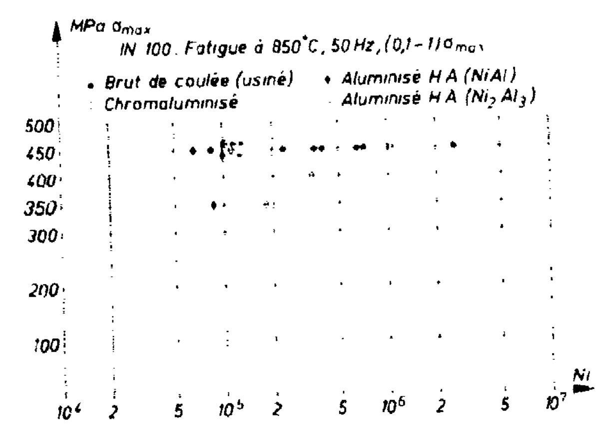

Fig. 12 – IN 100, fatigue HCF à 850° C

RECONDITIONNEMENT DE PIECES FIXES DU TURBINE PAR BRASAGE DIFFUSION

TURBINE STATOR PARTS REPAIR BY DIFFUSION BRAZING

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Le brasage-diffusion permet, lorsque les surfaces à assembler sont très proches (écartement maximal de l'ordre de 50 µm pour les superalliages), d'obtenir des zones de liaisons dont les propriétés sont du même ordre de grandeur que celles du matériau de base. Cette technique trouve donc une application de choix dans la réparation des fissures fines sur les aubes de turbine après fonctionnement. Cependant, comme certaines dégradations dépassent les écartements maximaux admissibles, une nouvelle technique a été développée. Ce procédé, dénommé "RBD" (Rechargement-Brasage-Diffusion), permet d'effectuer, à partir de poudres préalliées, des liaisons sans limites géométriques. La zone réparée est d'une composition chimique légèrement différente des pièces (à l'inverse du brasage-diffusion qui produit une liaison homogène) mais par un choix judicieux des paramètres opératoires du cycle de RBD, les propriétés locales peuvent être du même ordre que celles de la pièce massive.
Diverses applications illustrant les possibilités en réparation des deux techniques mentionnées ci-dessus sont présentées.

With very close surfaces (maximum gap of approximately 50 µm for superalloys), bonding areas obtained by diffusion brazing have the same properties as the parent material. This technique is successfully applied for repair of fine cracks affecting turbine vanes after operation. However, a new technique has been developed to eliminate those defects which exceed maximum tolerated gaps. The new process, called "RBD" (Rechargement-Brasage-Diffusion), allows bonding without geometrical limits from pre-alloyed powders. While diffusion brazing produces homogenous bonding, the chemical composition of the repaired area after RBD processing is slightly different from that of the part. But, rough part properties can be maintained in the bonding area, provided that operating parameters of the RBD cycle are adequately selected.
The two above mentioned techniques may have various repair applications which are presented.

PREAMBULE

L'accroissement des performances des turboréacteurs est en partie assujetti aux progrès de la metallurgie et notamment à la détermination de nouveaux alliages dont les limites d'utilisation sont sans cesse repoussées, mais dont la mise en oeuvre implique le développement de techniques adaptées en particulier pour ce qui concerne l'assemblage.

C'est ainsi que l'industrie des turbomachines utilise largement des matériaux tels que les superalliages de fonderie ou de métallurgie des poudres préalliées dont la "soudabilité", par les procédés conventionnels, est de plus en plus médiocre, certaines nuances comme le NK15CATu (IN 100) ou le NC14K8 (René 95) étant même réputées insoudables.

Aussi, des études ont-elles été menées par la SNECMA qui ont abouti à la mise au point de nouveaux procédés tels le Brasage-Diffusion (BD) et plus récemment le Rechargement-Brasage-Diffusion (RBD) qui, outre l'avantage d'être insensibles aux problèmes de fissuration, présentent un intérêt économique certain de par leur réalisation au four sous vide industriel, sans outillages spécifiques.

Ces techniques trouvent un terrain de choix dans le domaine de la réparation des pièces de turbine où la nature, la morphologie et la quantité des dégradations en fonctionnement sont très variées.

Nous nous proposons, après une présentation des procédés BD et RBD, d'illustrer par des exemples traités à la SNECMA les possibilités offertes par ces deux techniques.

1. PRESENTATION DU BRASAGE-DIFFUSION (BD) ET DU RECHARGEMENT-BRASAGE-DIFFUSION (RBD)

Le Brasage-Diffusion a été défini comme un procédé d'assemblage par un métal d'apport qui, par le jeu d'une phase liquide transitoire et de la diffusion intermétallique, permettrait d'obtenir des liaisons homogènes chimiquement dont les caractéristiques sont du même ordre que celles des matériaux à assembler.

Cette possibilité suppose un choix très précis des paramètres opératoires qui sont principalement :

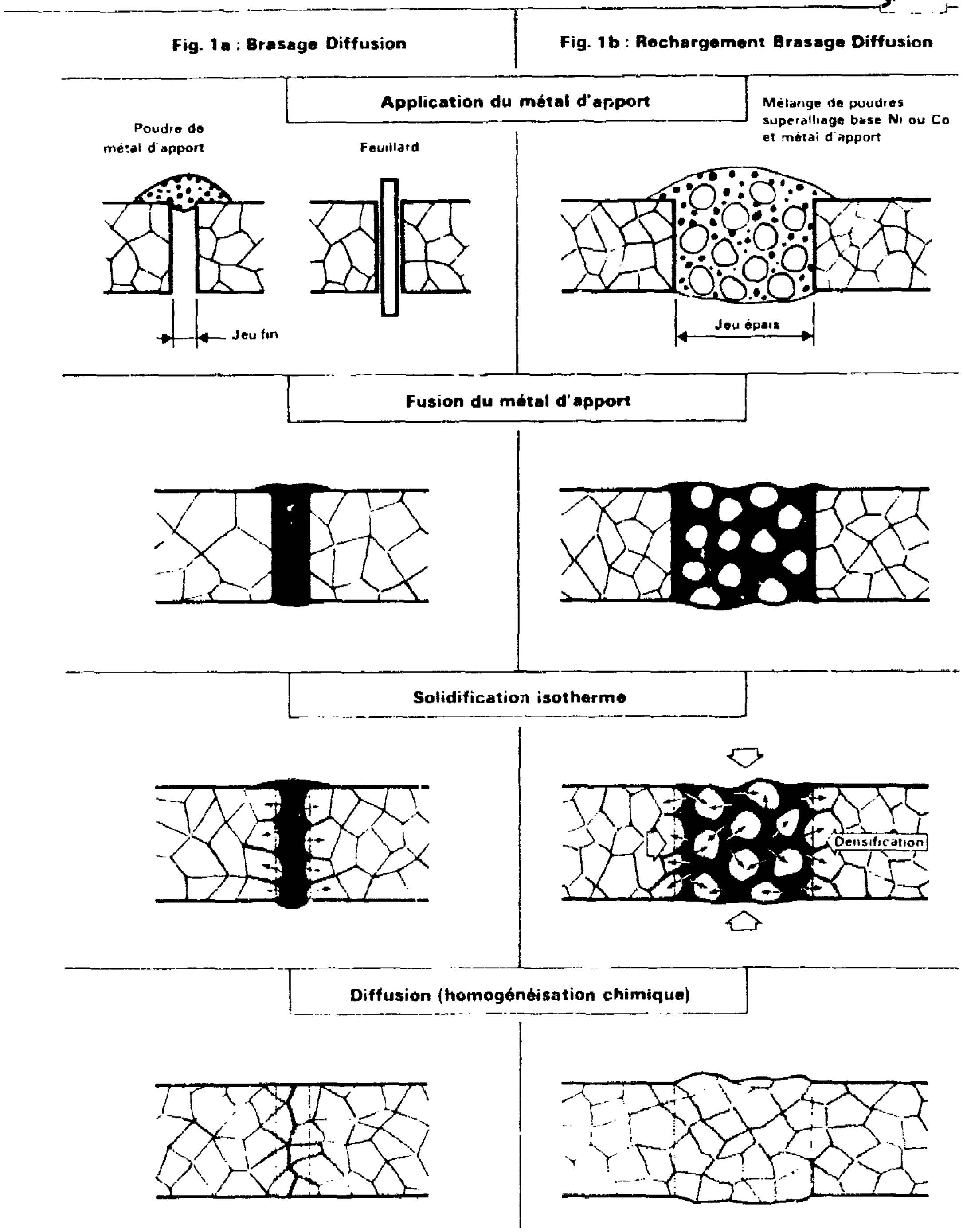
- la nature et la quantité du métal d'apport qui, outre son rôle de remplissage du joint, doit être compatible chimiquement avec le métal de base pour permettre une diffusion complète et réciproque des éléments en présence,

- le cycle thermique, moteur de la diffusion, qui doit être conçu de façon à éviter les formations de composés stables, les retassures ou les dégagements gazeux et permettre une homogénéisation aussi complète que nécessaire de la zone de liaison.

Sur le plan technologique, le métal d'apport peut être déposé sous forme de poudre ou introduit dans le joint sous forme d'un feuillard. La figure 1 a montré ces différents points et retrace schématiquement les différentes étapes du processus.

Fig. 1

**PRINCIPE DES PROCEDES BRASAGE-DIFFUSION (BD)
ET RECHARGEMENT BRASAGE-DIFFUSION (RBD)**



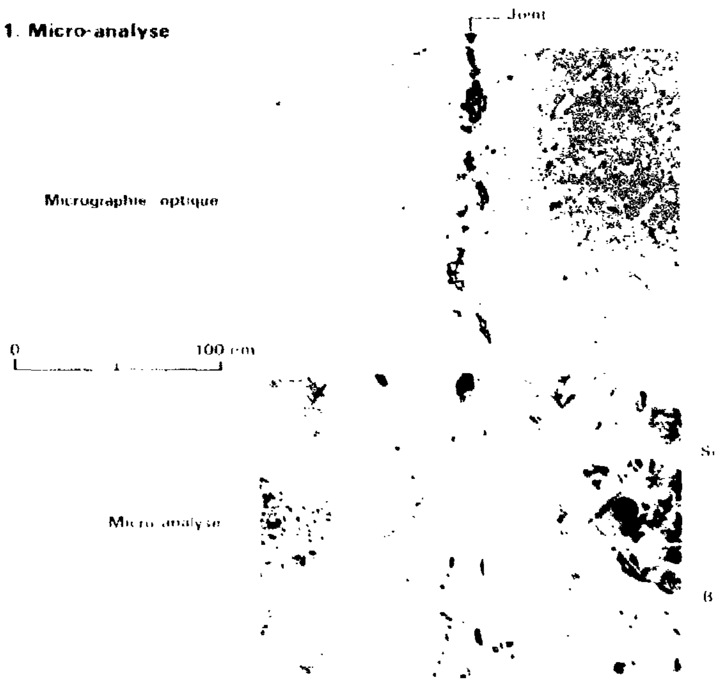
Pour les superalliages base nickel et cobalt, il a pu être montré que d'excellents résultats étaient obtenus en utilisant certains métaux d'apport fondés sur les systèmes Ni-Cr, Ni-Co, avec des additions de B et/ou de Si, dont la matrice est d'une composition voisine de celle des matériaux à assembler, et pourvu que le jeu entre les surfaces à assembler n'excède pas une valeur de 50 à 100 µm selon les alliages.

La figure 2 illustre un cas d'application obtenu sur un alliage NK15CATu (IN 100). Sur cet exemple, le brasage-diffusion a été réalisé au four sous vide à l'aide d'un métal d'apport du système Ni-Co-Si-B ; l'étude métallographique révèle que la diffusion est totale, résultat confirmé par l'essai de fluage rupture à 980°C qui montre une tenue des liaisons du même ordre que celle du matériau de base.

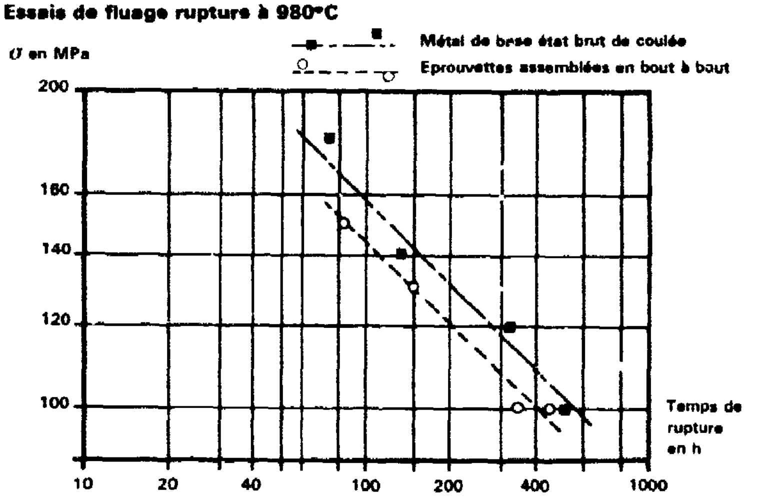
Fig. 2

BRASAGE-DIFFUSION DU NK15CATu (IN100) A L'AIDE D'UN METAL D'APPORT NiCoSiB

1. Micro-analyse



2. Essais de fluage rupture à 980°C



D'aussi bons résultats sont obtenus pour la majorité des superalliages de fonderie.

De par ses caractéristiques, le Brasage-Diffusion (BD) est donc bien adapté à la réparation des fissures fines qui prolifèrent souvent en réseaux denses : les phénomènes de mouillabilite et de capillarité permettent d'assurer un bouchage physique des défauts puis le travail métallurgique de la diffusion restaure en ces zones les propriétés du métal de base.

Cependant, la limitation dimensionnelle imposée dans le jeu des surfaces à assembler ne permet ni d'envisager la réparation de toutes les dégradations possibles, notamment sur une aube de turbine, ni de recharger les surfaces érodées. Aussi, une technique a-t-elle été développée par la SNECMA à partir du brasage-diffusion (BD), le "Rechargement-Brasage-Diffusion" ci-après dénommé RBD.

Dans son principe, le RBD (figure 1.b) consiste à densifier au cours d'un cycle thermique, un mélange de poudres préalliées de deux catégories distinctes :

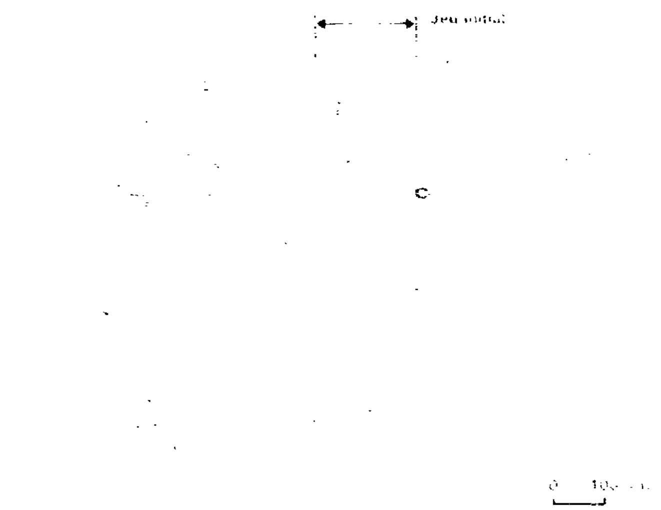
- l'une de composition analogue à celle d'un superalliage base nickel ou cobalt classique,
- l'autre de composition analogue à celles des matériaux de brasage-diffusion,

la composition chimique globale de l'ensemble étant identique ou très proche de celle du matériau constitutif des pièces traitées.

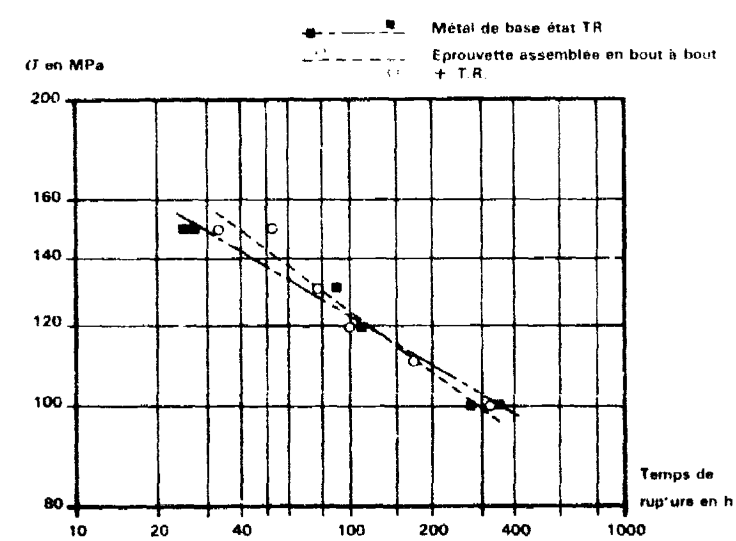
La diffusion des éléments abaissant le point de fusion des poudres de la deuxième catégorie se fait, au cours du cycle de traitement thermique, autant vers les surfaces de la pièce traitée que dans les grains de poudre de la première catégorie. Cet artifice permet, par rapport au BD, de raccourcir les parcours de diffusion et de conserver des temps de traitement raisonnables - et identiques - quel que soit le volume de matière à diffuser.

Fig. 3 **ASSEMBLAGE PAR RBD DU NK15CADT (René 77) A L'AIDE D'UN ALLIAGE RBD[NiCrB + NK17CDAT (Astroloy)]**

1. Aspect micrographique de la liaison



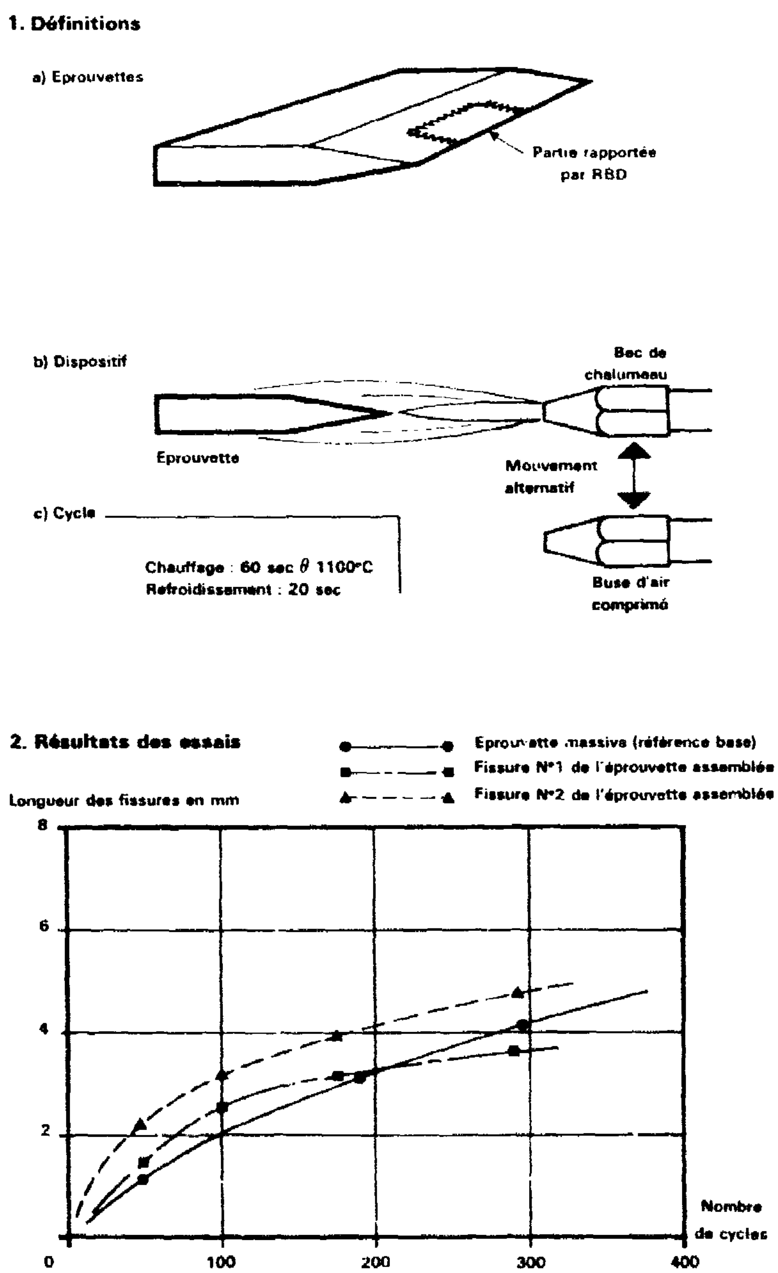
2. Essais de fluage rupture à 980°C



L'exemple de la figure 3 montre un assemblage réalisé sur NK15CADT (René 77) à l'aide d'un matériau RBD élaboré avec un métal d'apport NiCrB et d'une poudre NK17CDAT (Astroloy) conduisant à une composition finale proche de celle du René 77. Seule une variation structurale due au bore excédentaire peut y être remarquée, différence qui n'entraîne pas d'abattement en fluage-rupture à 980°C. L'essai de fatigue thermique (figure 4) d'éprouvettes prismatiques reconstituées par RBD ne fait pas apparaître d'accroissement de vitesse de fissuration dans les zones de liaison.

Il est possible d'obtenir des structures denses pratiquement sans limitations géométriques, soit dans des interstices entre des pièces à assembler, soit sur la surface externe d'une pièce. De telles structures assurent une continuité, dans tout son volume, des caractéristiques mécaniques et physiques de la pièce traitée.

Fig. 4 **ESSAI DE FATIGUE THERMIQUE SUR EPROUVETTES EN NK15CADT (René 77) RECONSTITUEES PAR RBD A L'AIDE D'UN ALLIAGE RBD[NiCrB + NK17 CDAT (Astroloy)]**



Grâce au jeu des paramètres opératoires qui régissent les propriétés physiques du mélange pendant l'opération de constitution, il est possible de bâtir des formes, combler des trous, recharger des surfaces...

Il y a cependant lieu de noter que lorsque le volume concerné par le matériau RBD devient trop important, c'est-à-dire au-delà de quelques dixièmes de millimètres, l'influence du matériau support s'estompe et la zone RBD présente alors sa structure propre qui est celle d'un matériau "métallurgie des poudres", grains équiaxes moyens à fins.

De plus, grâce à leurs principes identiques les procédés BD et RBD peuvent être combinés, c'est-à-dire que les cycles thermiques propres à chacun sont compatibles, peuvent se cumuler, se combiner...

Il apparaît alors possible, par une détermination précise et une combinaison judicieuse des opérations, d'aborder tous les types de réparation sur pièces de turbine, comme en témoignent quelques exemples traités à la SNECMA et présentés ci-après.

2. MODELES D'APPLICATION DES PROCEDES BD ET RBD EN FONCTION DE LA REPARATION ENVISAGEE

Les dégradations essentielles qui caractérisent une pièce de turbine détériorée en fonctionnement sont :

- d'ordre géométrique (fissure, arrachement par impact, érosion),
- d'ordre structural (corrosion, brûlure).

Devant des dégradations du premier type on procède généralement à un reconditionnement des surfaces de façon à assurer leur mouillage et à une liaison métallurgique, sans enlèvement de matière. Dans le second cas, on procède à une élimination des parties défectueuses et à leur remplacement. Cette technique de "rapiéçage" s'applique aussi de façon avantageuse lorsque les dommages de type géométrique sont trop étendus ou que la forme de la pièce à réparer est trop complexe.

Les exemples suivants sont destinés à illustrer les possibilités offertes.

2.1 Réparation directe de la pièce

Ce type de reprises est pratiquement limité à la réparation de fissures, au rechargement de petits manques de matière, de parties érodées.

Un premier exemple est celui d'aubes fixes de turbine, coulées de précision en alliage à base de cobalt KC25NW (HS 31), dégradées par fissuration ou fatigue thermique. Pour le reconditionnement de ces pièces, il a été nécessaire de procéder à un rechargement du bord de fuite en vue du ragrément du profil.

La réparation a été conduite selon la gamme de principe suivante :

- nettoyage des surfaces par une succession d'opérations classiques telles que décapage chimique et parachèvement par un traitement sous atmosphère contrôlée,
- remplissage des fissures les plus fines par un métal d'apport à base de NiCrB sous forme d'une pâte constituée de poudre préalliée et d'un liant volatile,
- fusion du métal d'apport et colmatage des fissures par passage à 1200°C 15 mn sous vide,
- dépôt d'un matériau de RBD constitué de poudre d'alliage base cobalt et de poudre de métal d'apport du type NiCoSiB,
- fusion du métal d'apport par passage à 1200°C 15 mn sous vide,
- contrôle de l'aspect géométrique des réparations,
- traitement de diffusion à 1200°C 4 h sous vide.

L'aspect d'une de ces pièces est montré sur la figure 5, à l'état brut de réparation. La microphotographie met en évidence, pour les zones rechargées, la continuité de la pièce à la réparation et la nature de celle-ci.

Un autre exemple est celui d'aubes de distributeur de turbine, en alliage à base de nickel NC22DK (C 242) présenté par la figure 6. Dans ce cas, des fissures assez importantes étaient présentes sur l'intrados et la réparation a été réalisée par une opération simultanée BD/RBD à laquelle un rechargement RBD a été adjoint pour la retouche du bord de fuite. Du point de vue métallurgie, on a obtenu :

- une parfaite continuité entre la réparation et la pièce,
- une légère différence entre les structures, le métal de base, un alliage de fonderie base nickel, présentant un grain plus fin que le rechargement.

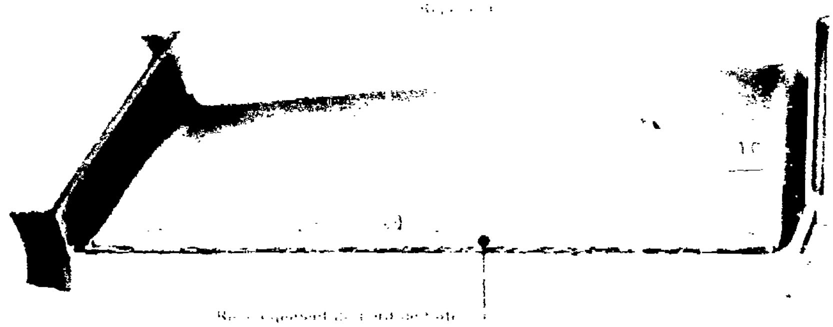
2.2 Changement de parties défectueuses

Il arrive, dans certains cas, que des dégradations trop importantes conduisent à opérer un changement total des parties défectueuses. Cette solution peut également être préférée lorsque les zones concernées sont trop ouvragées pour subir des retouches directes, comme par exemple un bord d'attaque perforé d'aube mobile.

On procède donc au découpage de la zone endommagée pour y substituer une partie neuve de même géométrie.

Fig. 5

REPARATION D'UNE AUBE FIXE DE TURBINE EN ALLIAGE DE FONDERIE KC25NW (HS31) PAR BD ET RBD



Examen sur coupe micrographique
du bord de fuite

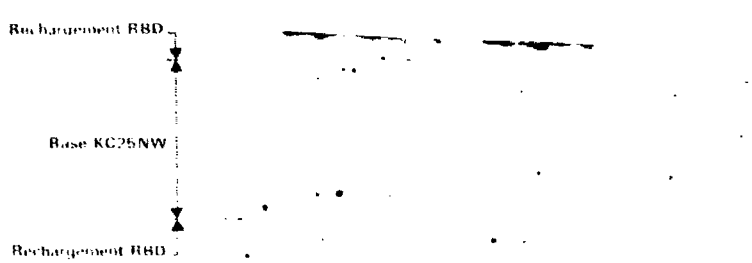
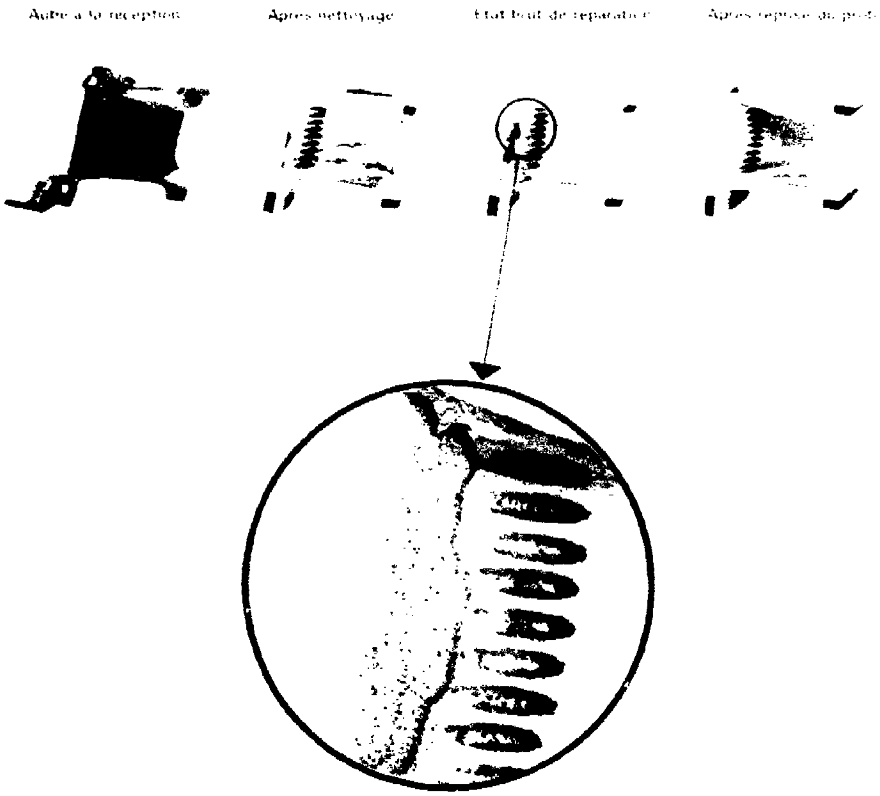


Fig 6

REPARATION D'UNE AUBE DE DISTRIBUTEUR DE TURBINE EN SUPERALLIAGE BASE NICKEL DE FONDERIE PAR BD ET RBD

Diverses etapes de la réparation



Aspect du RBD sur le bord de fuite

La réparation peut être abordée selon deux voies distinctes, en fonction de l'objectif visé :

1 - la recherche des caractéristiques mécaniques identiques à celles du métal de base, impliquant une parfaite continuité de la structure, nécessite le recours au brasage-diffusion, et un accostage aussi parfait que possible des surfaces à assembler ; ce choix sera par exemple préféré pour le changement d'un bord d'attaque ou de fuite sur une aube mobile de turbine où la qualité des liaisons doit être aussi bonne que possible.

   Compte tenu de la géométrie des pièces, on procède généralement à un usinage mécanique des surfaces d'accostage de façon à assurer des tolérances étroites.

2 - La reconstitution géométrique au moindre coût. On peut alors s'accomoder généralement d'une légère évolution locale des propriétés mécaniques et donc d'un accostage moins précis des surfaces à assembler. Le choix se porte alors sur une ablation des parties endommagées par usinage EDM au fil et un assemblage de l'insert par RBD.

   Cette deuxième voie est illustrée par l'exemple donné figure 7 du changement des bords d'attaque et de fuite sur une aube fixe de turbine en alliage à base de cobalt KC24NWTa (Mar M 509).

   Pour cette application, un montage permet de mettre en position les pièces, en vue d'un maintien par pointage ; le jeu relevé qui varie alors de 0,2 à 0,5 mm est comblé par l'alliage RBD constitué de poudre de NK17CDAT (Astroloy) et d'un métal d'apport NiCoSiB. Dans ce cas, il a été jugé préférable d'éviter l'utilisation de poudres d'alliage à base de cobalt susceptibles de former, lors du traitement thermique, des composés intermétalliques se comportant comme des barrières de diffusion. Le traitement de diffusion est effectué en four sous vide sans outillage particulier.

   Sur des secteurs multipales, il est possible que la dégradation de l'une des pales soit telle que son changement total devienne obligatoire.

   La récupération peut alors être envisagée selon le principe suivant : les secteurs réformés sont découpés en aubes élémentaires, par les plates-formes, à des cotes standards, par exemple sur une machine E.D.M. à l'aide d'un fil de 1 mm d'épaisseur ; les pièces unitaires sont ensuite soit rebutées, soit récupérées pour les éléments suffisamment sains. A partir de ces derniers de nouveaux secteurs sont construits par assemblage RBD. La préparation à l'assemblage consiste essentiellement à pointer les pièces sur un montage qui assure la section de passage dans la veine. Après remplissage des joints par un matériau RBD, l'assemblage lui-même est mené au four sous vide. Ce type d'opération peut être effectué en atelier après réintégration des pièces réformées. Une réparation identique peut être effectuée sur des pièces neuves rebutées en fonderie pour défauts géométriques.

   Un exemple de cette possibilité est montré par la figure 8 qui illustre la reconstitution de secteurs de distributeurs de turbine en alliage NK15CADT (René 77). Les photographies montrent le jeu entre les surfaces à assembler après le pointage, et l'aspect externe de la liaison RBD qui est suffisamment fin pour éviter toute retouche mécanique.

   Conjointement à l'assemblage, dans le cas des pièces réparées, certaines portées sont rechargées afin de permettre un usinage précis avant remontage.

   Dans le cas de dégradations importantes, il est aussi possible de reconstituer entièrement la partie dégradée à partir de poudres et l'exemple présenté à la figure 9 illustre cette possibilité. Il s'agit ici de la reconstitution d'un bossage de prise d'accessoire sur une aube fixe de turbine en NK15CADT (René 77). Dans ce cas, la proéminence initiale nécessitant une modification géométrique a été arasée et une nouvelle forme a été bâtie, directement en matériau RBD. Pour ce faire, une forme en poudre RBD a été préfrittée à une géométrie tenant compte du retrait ultérieur du matériau lors du cycle thermique . La forme de la prise d'accessoire est ensuite obtenue par usinage. L'intérêt d'une telle opération réside essentiellement dans son caractère économique dû à la facilité d'application ; la retouche est en effet appliquée directement sur la peau de fonderie après une préparation chimique, voire un simple dégraissage si la surface n'est pas oxydée. De nombreuses interventions de ce type sont possibles : bouchage de trous, reprises de nervures... pourvu que les caractéristiques recherchées aient pu être auparavant assurées par des essais en Laboratoire.

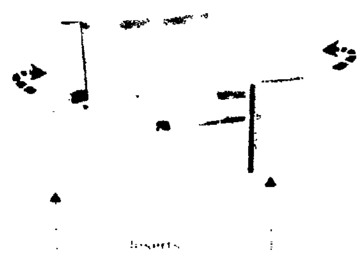
CONCLUSION

Le développement par la SNECMA des techniques de Brasage-Diffusion et de Rechargement-Brasage-Diffusion permet une nouvelle approche des problèmes de réparation. Les premières applications ont été motivées par la carence des procédés de soudage conventionnel notamment sur les superalliages de fonderie. Mais l'intérêt économique est vite apparu de la réparation de pièces d'autres types (chambres de combustion, carters...). Ces techniques étant mise en oeuvre à l'aide d'un four sous vide de traitement industriel et d'outillages simples, elles permettent le traitement simultané de nombreuses pièces.

Bien entendu, une expérience reste à acquérir qui permettra d'affiner les techniques et d'en obtenir un meilleur profit. Nous pensons par ces quelques lignes, avoir mis en évidence l'étendue des possibilités offertes quoique la liste des cas traités soit loin d'être exhaustive.

Fig 7 **CHANCEMENT DES BORDS D'ATTAQUE ET DE FUITE SUR UNE AUBE FIXE DE BANC EN KC24NWTa (Mar M509)**

Parties elementaires obtenues par decoupe au fil



Aspect apres assemblage

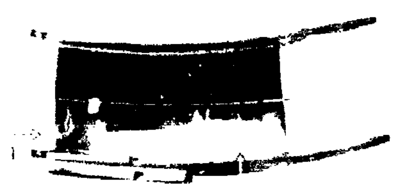
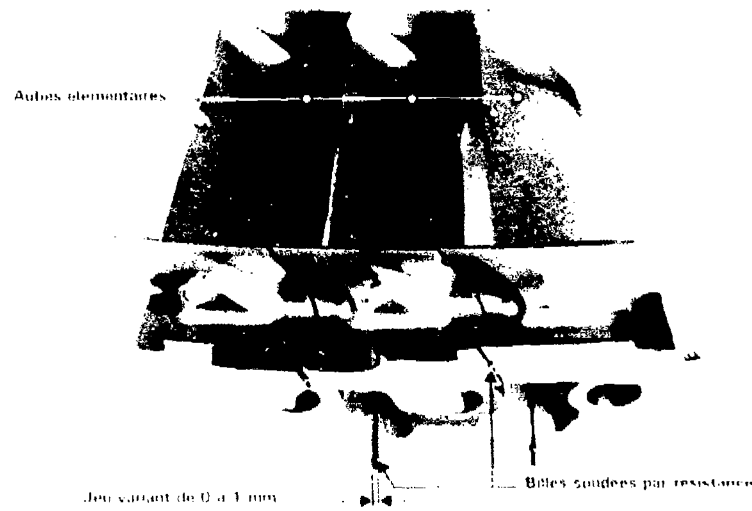


Fig 8

RECONSTITUTION D'UN SECTEUR DE TURBINE EN SUPERALLIAGE DE FONDERIE PAR ASSEMBLAGE BD ET RBD DE TROIS AUBES ELEMENTAIRES

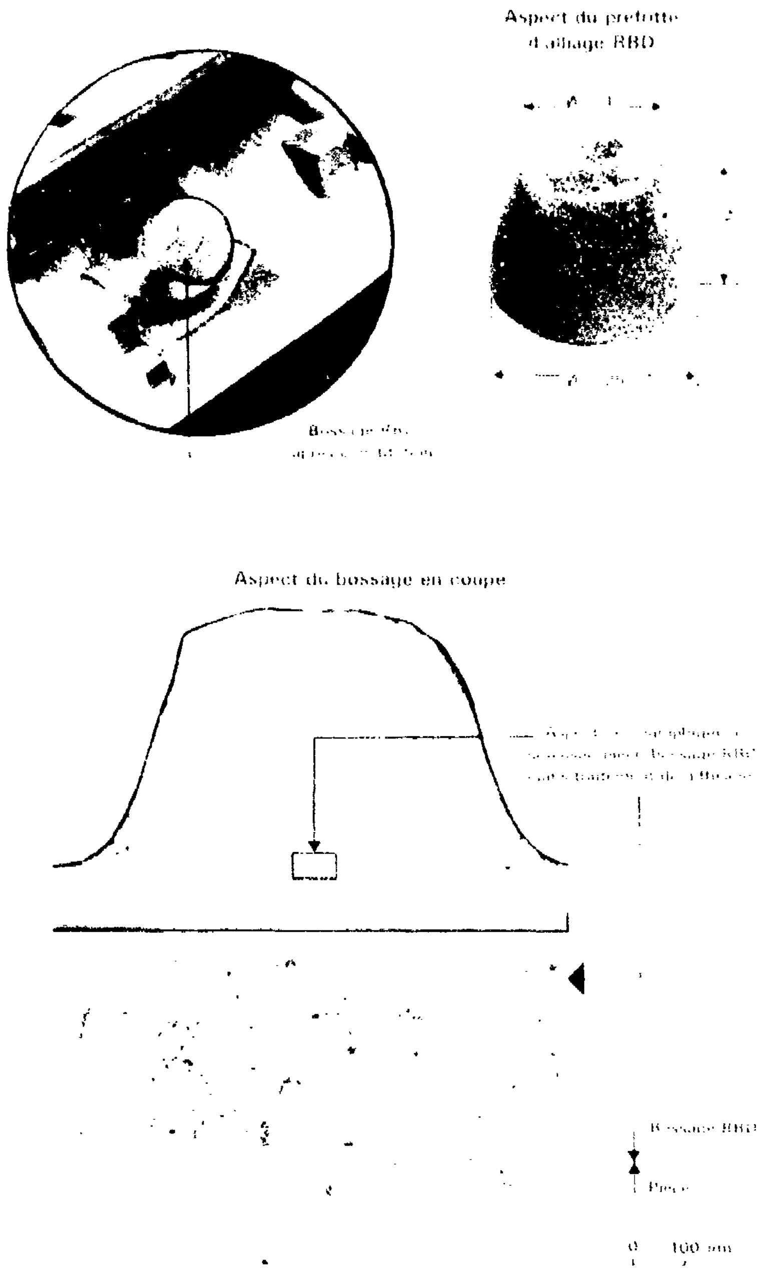
Pièces pointées en position avant assemblage



Aspect après assemblage du secteur reconstitué

Fig 9

IMPLANTATION D'UN BOSSAGE DE PRISE D'ACCESSOIRE SUR UNE AUBE FIXE DE TURBINE EN SUPERALLIAGE BASE NICKEL NK15CADT (René 77)



REJUVENATION OF USED TURBINE BLADES BY HOT ISOSTATIC PRESSING AND REHEAT TREATMENT

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SUMMARY

Gas Turbine blades operating at high temperature and stress experience microstructural transformations which eventually lead to their replacement. The occurrence of creep beyond allowable limits is a major cause of turbine blade replacement.

Microstructure and properties of many superalloys can be resotred by reheat treatment. This restoration is only partially effective if internal creep voids are present. Hot Isostatic Pressing (HIP) is a process which can be used to heal such creep voids and, in conjunction with the reheat treatment, provides a viable method for the rejuvenation of used turbine blades.

INTRODUCTION

The use of rejuvenation processes, usually heat treatments, to recover the properties of superalloy materials is a concept that has been with us for a number of years. However, this approach has received a new impetus, particularly for gas turbine blade materials, due to the current high cost and decreased availability of these components.

A major cause of blade replacement at overhaul is blade growth beyond the acceptable limits. Excessive blade growth, or creep, may be due to a number of mechanisms. Blades which experience a turbine overtemperature can undergo microstructural transformations which reduce the creep resistance of the material. Such blades can be expected to grow at a faster rate than those with normal microstructures. Similarly, blades with a long in-service life will experience this lessening of creep resistance, albeit at a more gradual rate. The normal blade tip clearances are quite small, and often as little as 1 or 2% creep elongation in the blades can cause tip rubs.

The degree of microstructural degradation found at overhaul may be acceptable, and in such instances the use of repair welding, blending, etc. can be used to return these blades to service. However, the blades will have lost some fraction of their useful life, and would benefit from a rejuvenation treatment. Further, if creep voiding has occurred, heat treatment alone will not heal these cavities. Here the additional factor of pressure, found in the hot isostatic pressing (HIP) process, is required.

There are several benefits from a successful rejuvenation process. The ability to return scrapped parts to service has obvious economic advantages. Reclamation of used blades decreases the need for spare parts, a definite gain in an era when suppliers must meet an increased demand for new components, to which the spare parts inventory perforce takes second place. Also, a reduction in the need for superalloy spare parts assists in the conservation of relatively scarce materials, e.g. cobalt, tantalum, etc.

The present programme is aimed at developing HIP rejuvenation re-heat treatment cycles for two Nimonic alloys, Nimonic 105 and Nimonic 115. The development of the cycles is in conjunction with the National Aeronautical Establishment, National Research Council, Canada.

REJUVENATION CRITERIA

Turbine blade life is limited by internal and external damage. The external damage includes hot corrosion, oxidation/erosion, thermal fatigue cracking and foreign object damage. The environmental effects are influenced strongly by the type of fuel used: land-based turbines utilizing clean natural gas suffer little or no corrosion attack, while similar units burning oil or similar fuels will experience corrosion early in their life, and must be monitored more frequently. Within limits, oxidation and corrosion products can be removed successfully by blending, but the repair of open cracks on the blade surface is not so reliable.

The internal damage which occurs in the nickel-based superalloy turbine blades may be broadly divided into two types. Firstly, microstructural damage results from the alteration of the basic structure due to high temperature thermal cycles. The nickel-based superalloys depend upon a fine dispersion of the gamma prime precipitate within the gamma solid solution matrix for their elevated temperature creep resistance. Thermal cycling results in the coarsening or agglomeration of the gamma prime phase, resulting in fewer obstacles to dislocation movement, with a corresponding reduction of creep resistance. Other undesirable effects include the breakdown of primary carbides and the formation of the topologically close-packed phases, namely sigma, Laves and mu.

These microstructural changes, either can be reversed or have their effect minimized for many alloys by re-heat treatment, or thermal rejuvenation cycles.

The other area of internal blade damage may be classified as structural discontinuities. These include cavitation, micro-cracks, and, in cast blades, inherent casting defects. Cavitation is the term applied to internal microporosity at grain boundaries with a transverse orientation to the direction of applied stress. These cavities can grow and link up to form micro-cracks, and give rise to a stress rupture failure. Micro-cracks may form also at such internal stress raisers as inclusions, or from the fracture of brittle carbides. Casting defects, of course, do not result from turbine operation, but are formed during solidification. Porosity, either due to entrapped gas or shrinkage, is the most common casting defect, but all such defects can have a deleterious effect on the operating life of the blade.

It was suggested earlier that the majority of microstructural defects were the result of elevated temperature, and may be reverted by heat treatment. Similarly, structural discontinuities, other than casting defects, are generally the result of applied stresses. These defects can be healed by the use of stress, the pressure of the HIP cycle. Therefore, a combination of HIP and heat treatment can be effective in reversing both classifications of internal blade damage, and restore not only the microstructure but also the mechanical properties of turbine blade materials.

HIP FACILITY

The Hot Isostatic Pressing process was developed originally for the consolidation of metal powders to attain theoretical density. The HIP unit is basically a pressure vessel, Figure 1, which is pressurized with an inert gas and simultaneously heated. The combination of heat and pressure collapses internal cavities and bonds surfaces together.

This concept was applied to investment cast turbine blades to heal the internal micro-porosity inherent in these components. This treatment proved very effective, and it is now standard practice at Westinghouse Canada Inc. to HIP as-cast blades in IN738 material before installation in turbines[1].

REJUVENATION PHILOSOPHY

Data accumulated on service exposed Inconel X-750 turbine blades with microscopic voids show that full restoration of mechanical properties is possible with a HIP re-heat treatment rejuvenation, Figure 2,[2]. Further, in-service testing of the treated blades shows no degradation of properties after 25,000 hours of post-rejuvenation life, Figure 3,[3].

Despite the relative simplicity of the HIP rejuvenation concept, early work[3] has shown that careful selection of cycle parameters is essential if the requisite microstructural features are to be maintained during the rejuvenation process. The need for the optimization of HIP and/or post-HIP thermal cycles for specific applications cannot be over-emphasized.

HIP CYCLE

The HIP temperature selected for a given alloy is above the gamma prime solvus temperature for the alloy. Such a temperature allows for the dissolution of the gamma prime phase, and ensures that material flows readily under the isostatic pressure imposed to seal the internal cavities. Other factors must be considered also for the exact temperature to be chosen[4]. These include the grain coarsening temperature of the alloy, which may be time dependent also, Figure 4. Another consideration is the solvus temperatures of precipitates other than gamma prime. The ideal selection would include examination of virgin alloy material to determine the degree of primary carbide degeneration that has occurred during service exposure. Further, much rejuvenation work is naturally carried out on older materials, where perhaps the composition was not as strictly controlled as in later alloys. The examination of virgin stock could determine the presence of sigma phase, for example, and how much, if any, was tolerated in the original blades.

RE-HEAT TREATMENT CYCLE

The aim of the post-HIP reheat treatment cycle is to restore the microstructure of the alloy to its pre-service condition. Some alloys require a slow furnace cool from the solution treatment temperature to either a lower ageing temperature, or a partial solution treatment temperature. Such a treatment exists for Nimonic 115[5,6], and a similar treatment is used for HIP processed IN-738 castings[7]. The treatments give improved creep ductility over the conventional air cooling treatments. This is attributed to the formation of serrated grain boundaries, that are believed to resist sliding and give rise to more homogeneous deformation.

The slow cooling rate can provide desirable grain and twin boundary morphology, and by allowing longer times for MC and $M_{23}C_6$ particle growth, could cause the formation of large, discrete carbides at these boundaries. Such larger carbides could improve creep rupture properties by disrupting grain boundary sliding.

The adverse effects of the slow cooling can be the formation of carbide platelets at the grain boundaries, and a larger gamma prime particle size, with a resultant increased interparticle spacing for a given volume fraction of gamma prime precipitates. Such particle size changes tend to increase the creep rate, Figure 5, as a result of easier Orowan looping and diffusion controlled grain boundary migration[8].

Thus the selection of the post-HIP reheat treatment cycle must consider:

a) serrated grain boundary formation
b) carbide precipitate morphology
c) size and shape of gamma prime particles.

The microstructural features are interrelated, and the optimum results are obtained by control of solution temperature, cooling rate, partial solution temperature and ageing treatment.

CURRENT DEVELOPMENT

HIP reheat treatment rejuvenation cycles are being developed for Nimonic 105 and Nimonic 115 service exposed turbine blades. The blades had experienced approximately 100,000 hours of exposure, and the as-received materials showed a corresponding degradation of physical properties. This was particularly noticeable in the Nimonic 115 blades, where heavy sigma phase formation had caused severe embrittlement.

Initial rejuvenation trials have resulted in the restoration of tensile and stress rupture properties but full ductility has not been recovered. However, further cycles involving slow cooling techniques are being studied, and a full restoration of all properties is expected.

The second stage of this programme will be the rejuvenation of a complete row of service exposed blades of each alloy. The treated blades will then be returned to their respective turbines, and their in-service performance measured on a regular basis. This monitoring is essential to prove the long term effectiveness of the rejuvenation treatments.

CONCLUSIONS

The microstructural philosophy necessary for the successful HIP reheat treatment rejuvenation of service exposed superalloys has been developed, and Westinghouse Canada Inc. has applied this philosophy to used turbine blades.

The in-service performance of rejuvenated parts has been monitored, with very good correlation of properties between rejuvenated and new components. Existing results suggest that HIP reheat treated blades are a viable alternative to new replacement blades for gas turbines.

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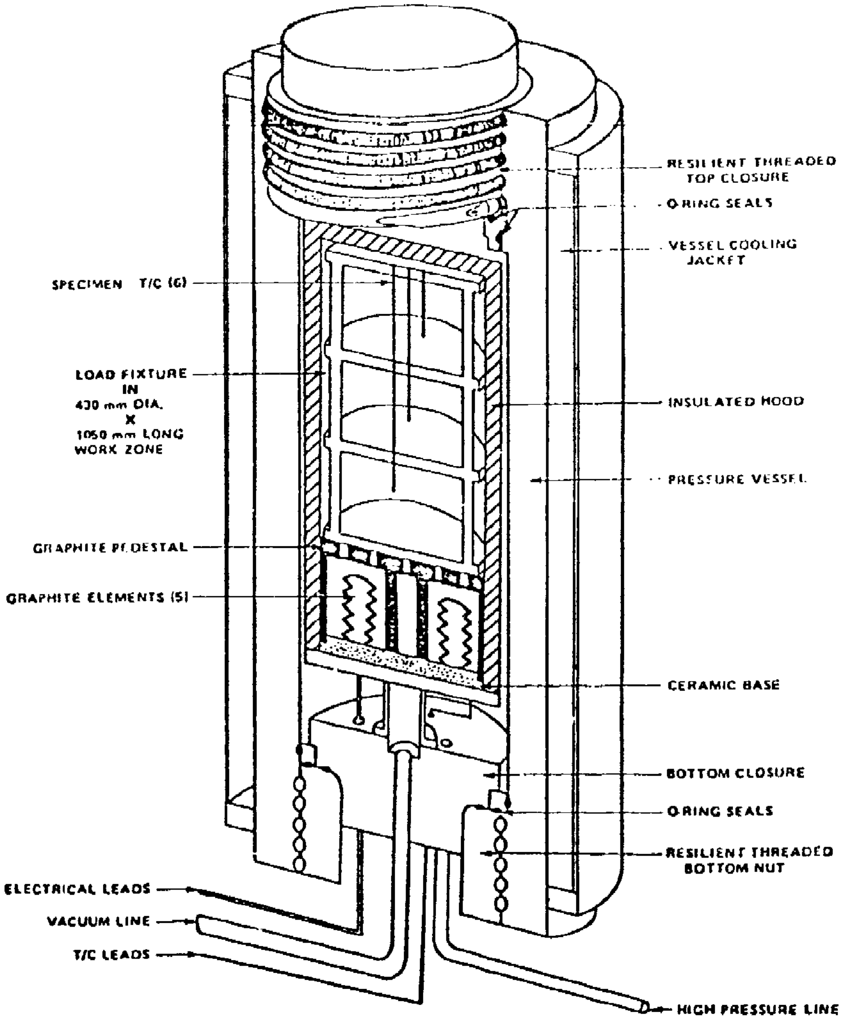

FIG.1. SCHEMATIC OF THE WESTINGHOUSE CANADA HIP VESSEL

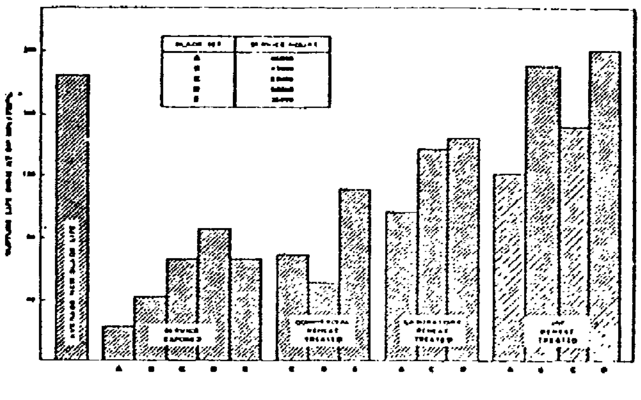

FIG.2. STRESS-RUPTURE PROPERTIES FROM SEVERAL SETS OF INCONEL ALLOY X-750 BLADES AFTER VARIOUS TREATMENTS COMPARED TO AVERAGE NEW BLADE LIFE

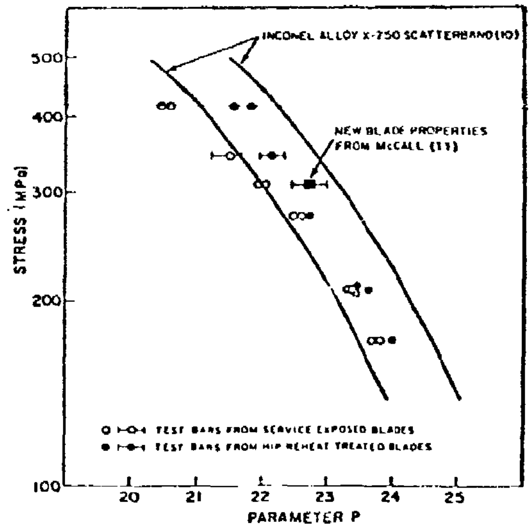

FIG.3. LARSON-MILLER PLOT ($P = T \times 10^3 (20 + \log t)$, T IN °K, t IN HOURS) COMPARING STRESS-RUPTURE PROPERTIES OF SERVICE-EXPOSED, VIRGIN AND HIP PROCESSED BLADES

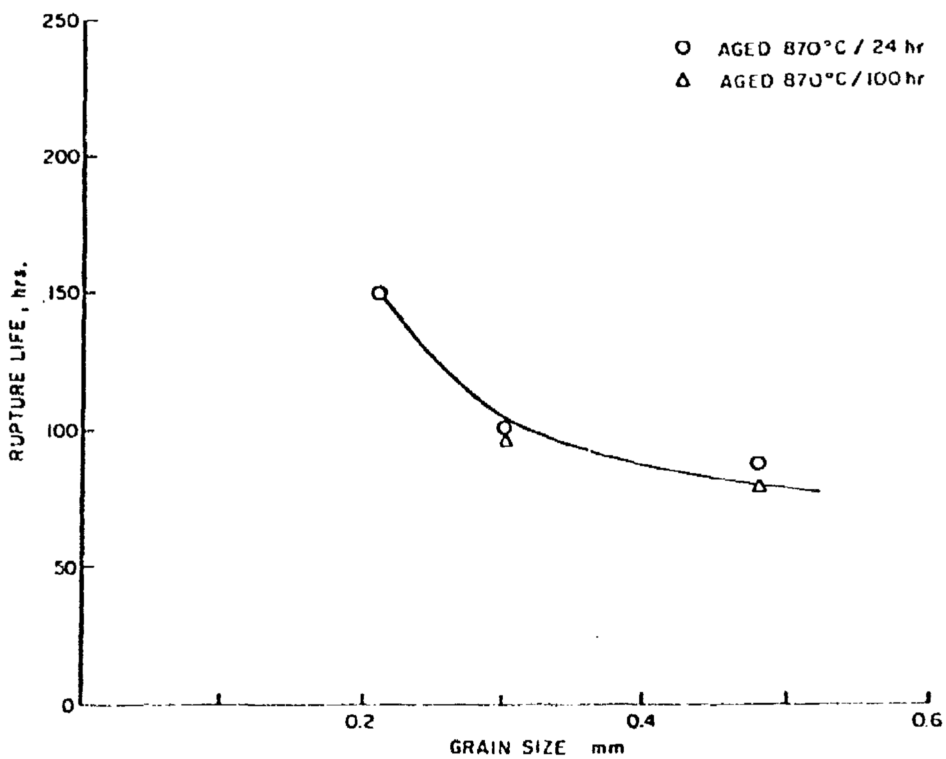

FIG.4. MICROSTRUCTURAL DEPENDENCE OF RUPTURE LIFE OF SERVICE SIMULATED INCONEL 700 TESTED AT 790°C AT 345 MN/m²

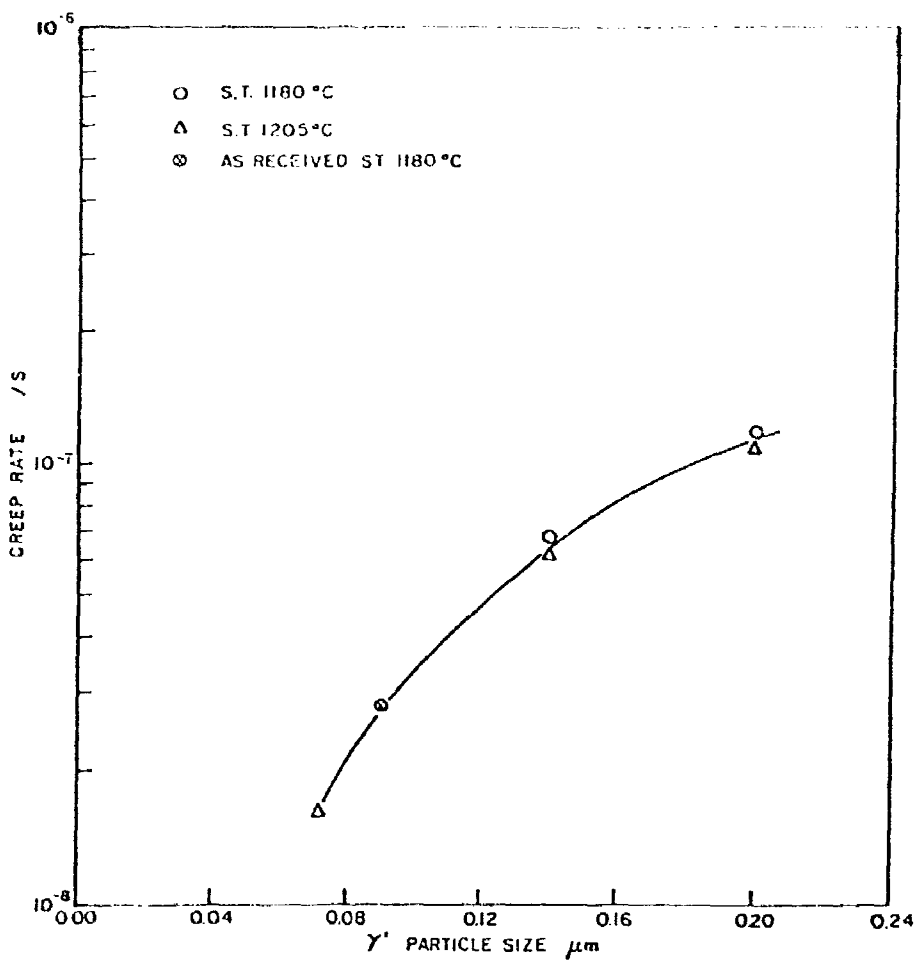

FIG. 5. γ' PARTICLE SIZE DEPENDENCE OF CREEP RATE OF SERVICE SIMULATED INCONEL 700

HIP PROCESSING -

Potentials and Applications

by

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SUMMARY

Introduced in the early 1960s as a sophisticated metalworking and heat treat process, hot isostatic pressing has made strong inroads against established techniques. Hot isostatic pressing, or HIP, is a process in which components are subjected to the simultaneous application of heat and pressure in an inert gas medium. From its inception as a means of gas pressure bonding nuclear fuel assemblies (1) HIP processing today finds a wide variety of uses from cermet cutting tool manufacture, to the production and rejuvenation of gas turbine hardware. (2)

The results obtained on complete engine sets of Inconel X-750, Udimet 500 and Rene' 100 turbine blades indicate that HIP processing is capable of restoring new or near new creep properties and low cycle fatigue properties to used blades.

The process should be applicable to most superalloys, nevertheless, recent work emphasized the need for preproduction process parameter verification to establish optimum cycles which take into account both the metallurgical and mechanical aspects of superalloys.

1. DESCRIPTION OF CHROMALLOY'S HIP UNIT

The basic equipment needed for HIP processing consists of a pressure vessel, a high temperature furnace, a gas compressor, and a temperature pressure control system. Advanced HIP pressure vessels must withstand pressures up to 30,000 psi (2068 bars), furnaces must operate up to 2730°F(1500°C), the gas compressor must not contaminate the working fluid, and the control system must maintain work zone temperature to within ± 25°F(± 14°C), and pressure within 2%.

Figure 1 illustrates the latest generation of equipment, and Figure 2 shows the actual Chromalloy HIP unit.

In Table 1 the basic parameters for the Chromalloy HIP unit are given.

The Chromalloy HIP facility will accept parts 18"D x 20"L for processing at temperatures from 1400°F(760°C) to 2400°F(1315°C) at pressures from 5,000 to 29,000 psi.

The unit offers complete thermal cycle capability including rapid quenching. As will be discussed later, this feature is particularly useful for engine run turbine hardware which requires post-HIP heat treatments for optimum property recovery. Use of high purity liquid argon as a medium for pressurization minimizes the potential for surface contamination. Temperature of the work pieces is precisely controlled and monitored by multiple thermocouples distributed throughout the area.

2. PRINCIPLE OF HOT ISOSTATIC PRESSING

As shown in Figure 2A, a part placed within the work zone is simultaneously acted upon by heat and pressure. Internal voids, sealed from the surface, experience a pressure differential exceeding the local yield strength of the material at appropriate operating temperatures.

Deformation occurs in this region as the internal free surface of the void is pressed shut. Metallurgical bonds are formed soon after void collapse. Once at temperature and

pressure, healing occurs rather rapidly.

Need for more cost effective manufacturing to produce wrought superalloy and titanium products has brought forth the use of HIP to manufacture near net shape forging preforms. (3,4,5)

Most major engine manufacturers foresee HIP incorporated on a large scale into turbine disk manufacture. In fact, based on results from HIP powder consolidation experiments, the U. S. Air Force is funding several manufacturing programs on airframe and engine components which will use hot isostatic pressing to save time, conserve material and produce better products.

In the mid 1960s, researchers reasoned that if HIP densifies sealed or presintered wrought material and imparts superior mechanical properties, then HIP should favorably affect castings with subsurface porosity, or shrinkage cavities. Provided the microporosity is well distributed, no measurable dimensional changes due to pore closure should occur during HIP. Preliminary work by GE(6), Battelle(1), and Howmet(7) confirmed that, indeed, this was the case.

Application to the aerospace investment casting field was obvious, for as turbine needs have led to stronger, more complex alloys and shapes, casting quality did not keep up with these demands.

Chromalloy Division-Oklahoma has used the HIP unit especially for the aerospace industry in developing repairs and rejuvenating jet engine parts which were considered unserviceable a few years ago.

3. RESULTS OF HIP REJUVENATION OF TURBINE COMPONENTS

Reduction in microporosity has a profound influence on the mechanical properties of cast superalloy components.

Representative data (Figures 3, 4 and 5) demonstrate that for stress rupture type applications, HIP processing can improve mean property levels and decrease scatter bands. (6,7). At temperatures above the notch sensitive range, typically 1600°F(870°C) and above, HIP has less effect on the mean property level, but by reducing the porosity, HIP reduces the scatter. Obviously, the thinner the cross section the greater the potential benefits of HIP.

Fatigue properties can be improved by HIP processing as well. Recent data on B1900 and Nimonic 105 alloy turbine blades are presented in Figures 5 and 6. Note that both mean and 98% confidence limit are improved through use of HIP.

Engine manufacturers have several ways to capitalize on these developments. First, since HIP reduces property scatter more confidence may be placed in the design, either by raising allowable stress levels or extending overhaul periods. Second, the elimination of porosity greatly reduces the chance for premature failure. Third, by eliminating subsurface defects less rejections will occur during final machining operations as these areas become the external surface.

Successful application of HIP on castings depends on optimally combining the casting parameters, HIP parameters, and subsequent heat treatments, to bring forth those characteristics most desirable for the given application.

Use of HIP to rejuvenate damaged components, or extend the service of life limited components, is an area of keen interest to the Chromalloy organization.

Theoretically, HIP should heal internal damage generated during service; surface microcracks may be healed provided a suitable coating is first applied. Immediate engine applications would include turbine blades and disks from both the turbine and compressor.

At Chromalloy Division-Oklahoma, numerous programs employing HIP recovery of mechanical properties of engine run components have produced positive results. One program demonstrated property recovery of engine run B1900 blades through use of typical HIP process conditions followed by resolution heat treatment; HIP at 2175°F(1190°C) 27,000 psi for 2 hours plus resolution at 2175°F(1190°C), 2 hours rapid cool plus coating cycle at 1975°F(1079°C) for 4 hours plus age at 1650°F(900°C) for 10 hours.

Before and after HIP mechanical properties and microstructures, illustrated in Figures 6 and 7, demonstrate that HIP will improve the stress rupture life and close voids in service run hardware, produce some internal structure homogenization and favorably influence carbide formation and distribution.

Stress rupture properties of the engine run B1900 parts fall on the low end of the new part scatter band. Subjecting the used parts to HIP processing, fully restored stress rupture properties. Later work with Chromalloy's advanced HIP unit, having rapid cooling capability, produced equivalent results without the need for a resolution treatment. A similar effort undertaken to recover Rene' 100 and SEL-15 alloy turbine blades has resulted in a production process for rejuvenation of turbine blades now used by a major airline and the general aviation fleet.

The full procedure includes a tip weld to restore length, a hot reform to reset twist and warp angles of the airfoil, and hot isostatic pressing to insure that the overhauled part has optimum stress rupture properties.

Initially, the Rene' 100 blades were HIP processed in units without heating or cooling flexibility and received post-resolution treatment as part of the process. The Chromalloy unit, offering rapid quench capability, allowed elimination of this heat treatment with no loss in properties. In fact, significant improvement in 1400°F(760°C) rupture characteristics are observed.

The importance of thermal cycle flexibility in the HIP cycle is illustrated by results of the program to improve the SEL-15 version of the blade. Conventional HIP followed by post-HIP resolution treatments caused severe degradation of the alloy properties while a modified cycle in a unit using rapid quenching improved the alloy properties (Figures 8 and 9).

Thus far, we have indicated benefits of HIP in recovery of stress rupture properties. Similar benefits can be gained in fatigue. Figure 10 illustrates a comparison of HIP versus no HIP on a cast SEL alloy component. The parts not HIP processed were given a conventional solution treatment in an effort to recover mechanical properties. It is evident that a dramatic improvement of fatigue properties could be achieved by using HIP rather than conventional heat treatment.

Data on Rene' 80 alloy turbine blades reinforce the notion of correctly tailoring the process cycle to the alloy and component. Chromalloy conducted a number of cycles aimed at optimizing the process cycle. HIP followed by normal coating and aging heat treatments optimized stress rupture properties.

Further improvement was not possible through resolution treatment after HIP and caused deterioration of properties (Figures 11 and 12). Recently, Chromalloy studied the feasibility of incorporating HIP rejuvenation in the repair scheme of low pressure first stage turbine blades made from Nimonic 105 material. The total engine time on the blades was 1,290 hours. Three material conditions were studied:

(i) "As received" to document the degradation of material properties;

(ii) Simulated Conventional Repair Treatment (CRT) representing normal repair procedures;

(iii) HIP rejuvenation.

Seventeen blades were used for each of the three material conditions. For the "as received" condition, no additional treatment was given and the blades were tested as such. The simulated CRT consisted of 1900°F(1038°C) for 0.5 hours, air cool followed by 1290°F(700°C) for 16 hours and air cool. The third group of samples was HIP'ed: 1975°F(1080°C) for 4 hours at 28 ksi followed by a rapid cool and a heat treatment at 1560°F(850°C) for 24 hours and 1300°F(700°C) for 16 hours. Three blades were used for elevated temperature tensile testing at 1200°F(650°C) and stress rupture testing at 1200°F(650°C), 118.7 ksi was performed on seven blade test specimen for each condition. The configuration of test specimens for both testings is shown in Figure 13.

Fatigue testing was performed at 1000°F(538°C) and 150 ksi stress level on seven test specimens for each condition.

The blades used for stress rupture testing were also employed for metallographic examination. As a result of these tests we found that the tensile properties at 1200°F (600°C) were comparably similar for "as received" and HIP'ed conditions. The CRT material showed lower strength levels, but better ductility. However, for all three conditions, tensile properties were within the alloy specifications.

The stress rupture data is presented in Figure 14. The data show good stress rupture capability for all three conditions (Alloy Digest data at 650°C, 118.7 ksi, indicates a 50-hours life). CRT material and HIP'ed blades show some marginal improvement in the stress rupture life over "as received" condition. HIP'ed blades show a wider spread in stress rupture lives although the spread in data is evident for all the three conditions.

The fatigue results (Figure 15) show a distinct improvement for the HIP'ed condition; both log mean and 98% limit cyclic lives are superior to those exhibited by "as received" and CRT material. The "as received" material shows a definite degradation in fatigue performance at the engine run blades.

As a result of these tests, HIP was incorporated in the repair scheme and the blades, life limited before, are used again for at least one more overhaul cycle.

Encouraged by these results, Chromalloy is working toward applying HIP for recovery to other wrought alloys including Nimonic 80A, Inco 901 and 12% chromium steel. We feel that directionally solidified alloys will respond to HIP process treatments as well. Optimization can be achieved either through resolutionizing after HIP or by direct quenching from the HIP temperature.

For overhaul work Chromalloy recommends a direct quench method to avoid potential deleterious effects due to remnant aluminide coating which may occur during a conventional post-HIP resolution treatment. Stress rupture data at 1800°F(982°C) and fatigue data at 1400°F(760°C) illustrate potential improvements for HIP of D.S. hardware (Figure 16). Extending these thermomechanical techniques to other jet engine hardware, such as disks, will require developing further techniques in order to maintain dimensional tolerances and to heal surface related defects.

The preceding examples, however, emphasize several points.

First, HIP can be used to recover life limited components at overhaul if damage has not progressed to the point of incipient failure. Second, HIP can be used to optimize both new and used part properties. Third, both cast and wrought components will respond to HIP processing. Fourth, further studies are required in order to optimize HIP cycles for those materials that do not respond favorably to HIP treatment using the parameters established for conventional heat treatment procedures.

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HIP Equipment Schematic

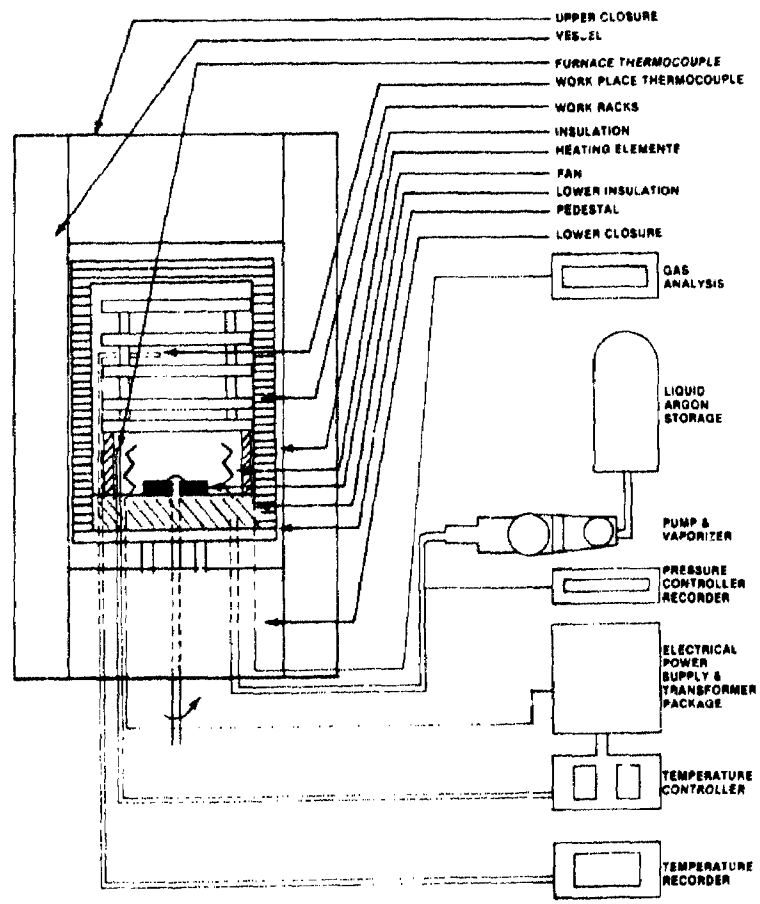


Fig. 1

Fig. 2 Chromalloy HIP Facility

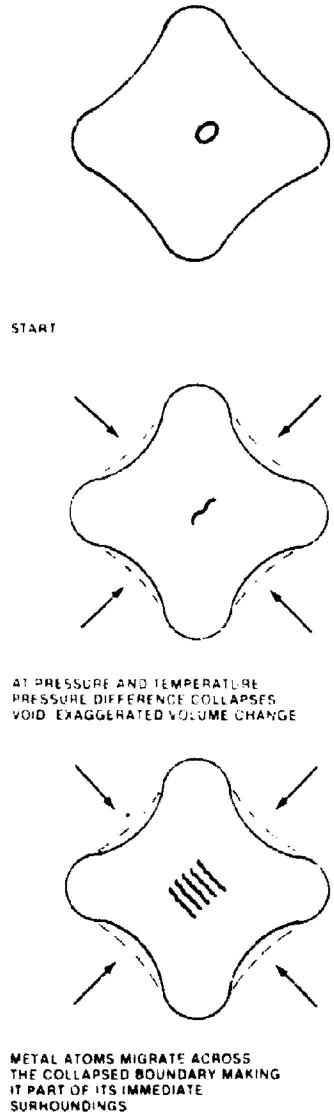


Fig. 2A **Principle of Hot Isostatic Pressing**

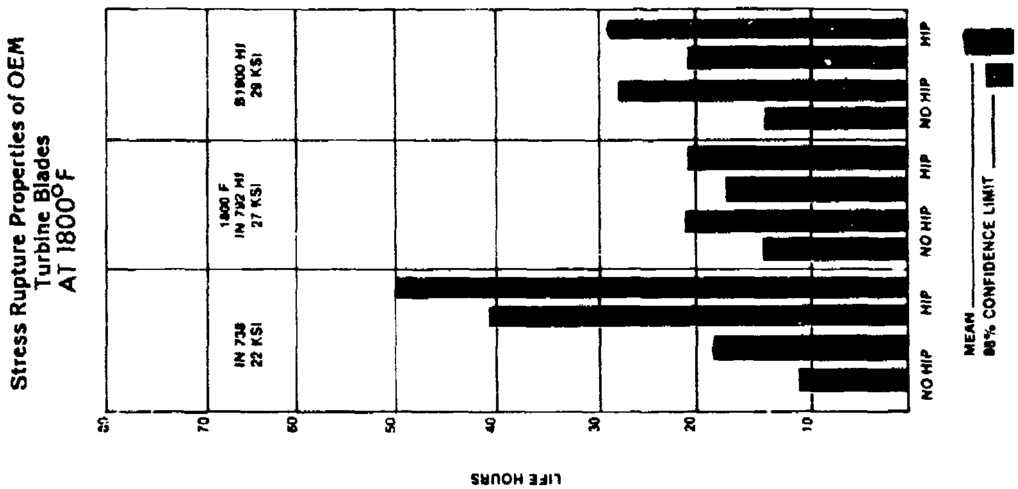

Stress Rupture Properties of OEM
Turbine Blades
AT 1800°F
IN 738
22 KSI
1800 F
IN 792 HI
27 KSI
B1900 HI
29 KSI
NO HIP
HIP
MEAN
98% CONFIDENCE LIMIT
LIFE HOURS

Fig. 4

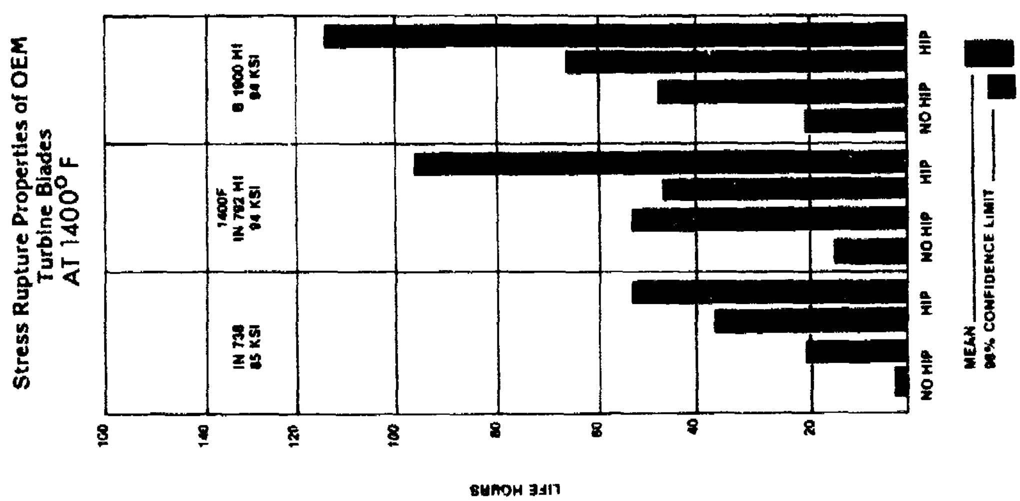
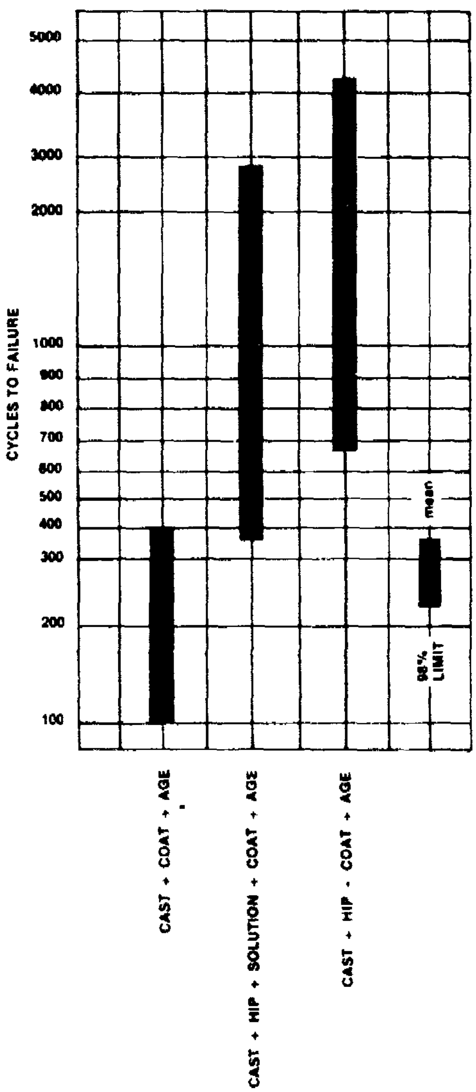

Stress Rupture Properties of OEM
Turbine Blades
AT 1400°F
IN 738
85 KSI
1400F
IN 792 HI
94 KSI
B 1900 HI
94 KSI
NO HIP
HIP
MEAN
98% CONFIDENCE LIMIT
LIFE HOURS

Fig. 3

Effect of HIP on 1400F Fatigue Properties of a cast B1900 Turbine Blade



K_t = 1.0
A = 0.95
f = .33 H_z
MAX. STRESS = 110KSI

Fig. 5

Effect of HIP on Mechanical Properties of Service Run B1900 Turbine Blades.

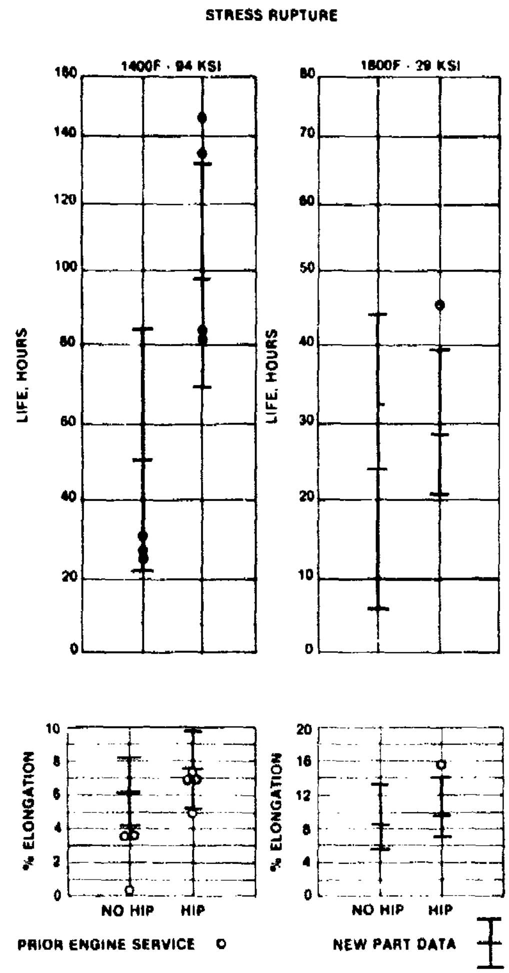


Fig. 6

A. Simulated Coat & Age 200 X

B. Simulated Coat & Age 1000 X

C. Hip & Coat + Age 200 X

D. Hip + Coat + Age 990 X

Fig. 7 Cast B. 1900 Turbine Blade Micro Structure Before And After Hip

1400F / 85 KSI Stress Rupture Performance

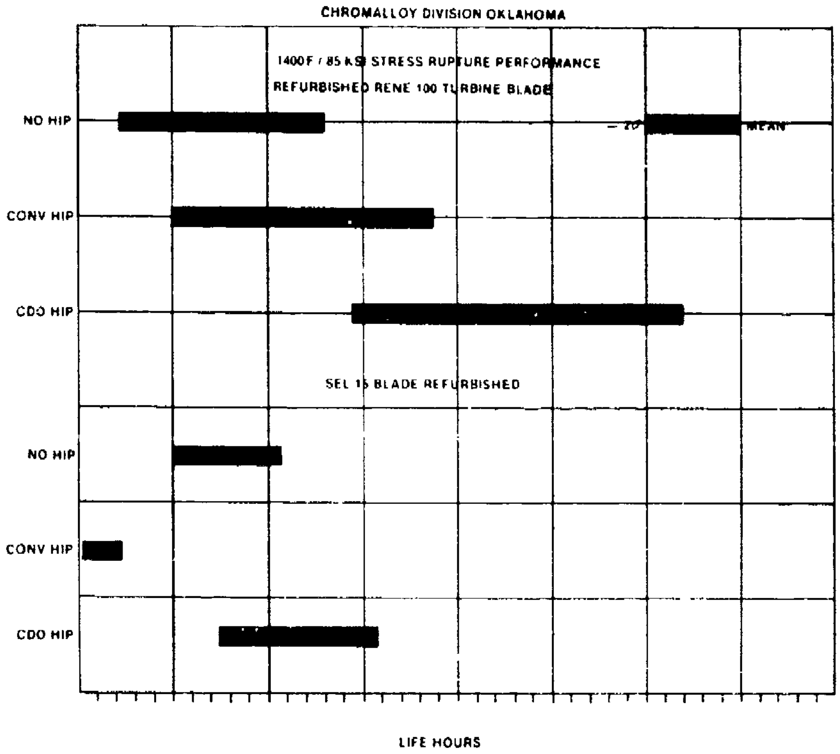


Fig. 9

1400 F / 85 KSI Stress Rupture Performance

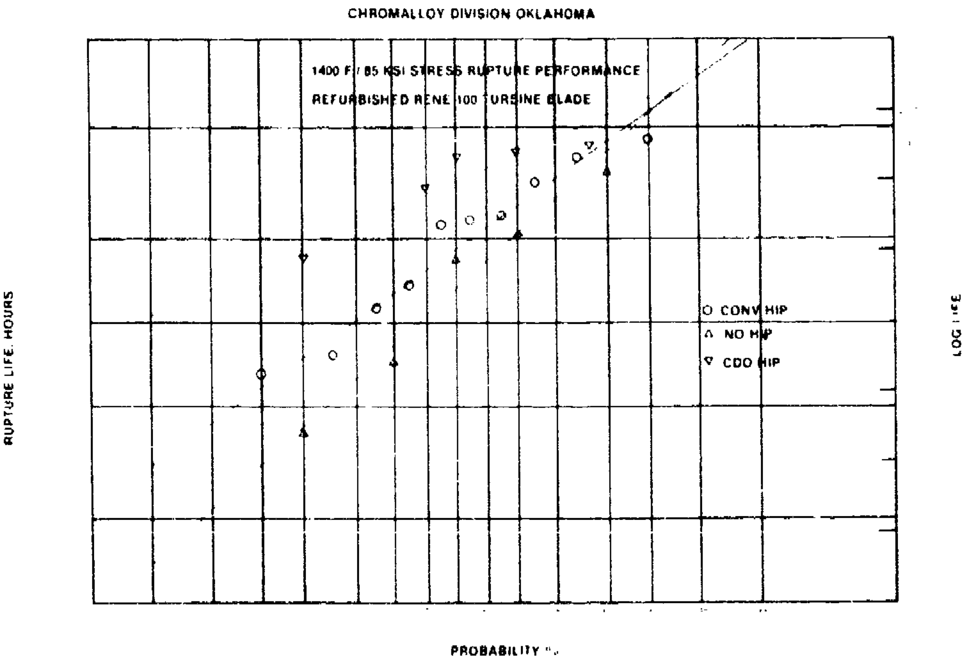


Fig. 8

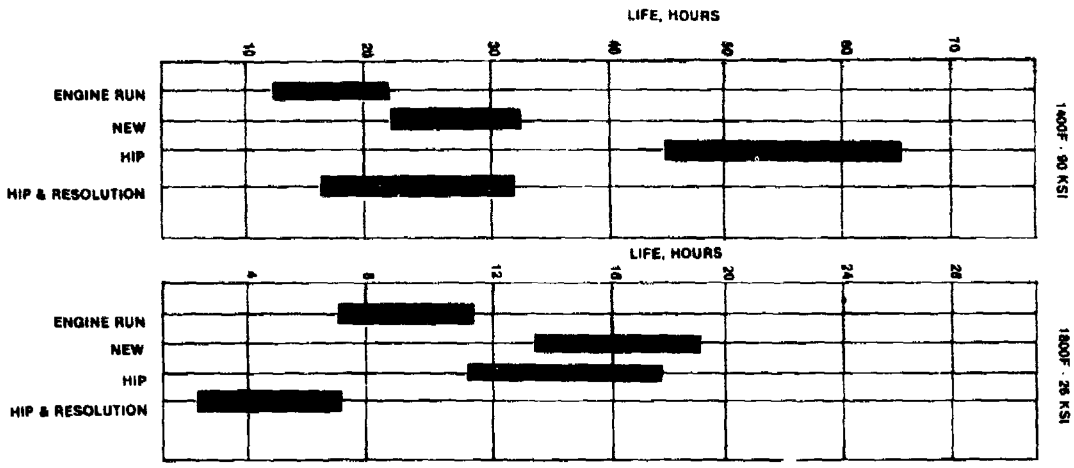


Fig. 11 RENE 80 Turbine Blade Stress Rupture Data

Cast SEL-Turbine Component
1000F Fatigue

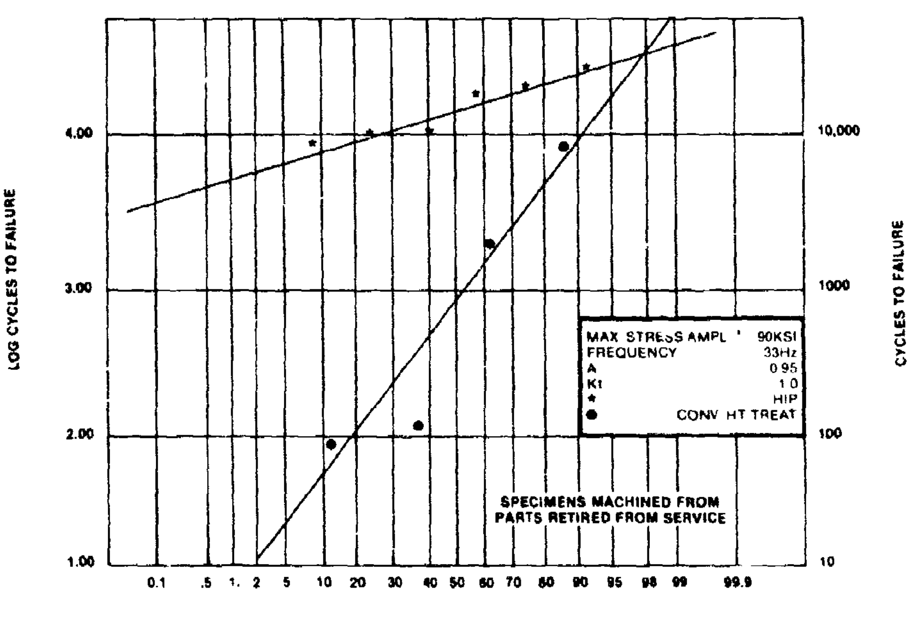
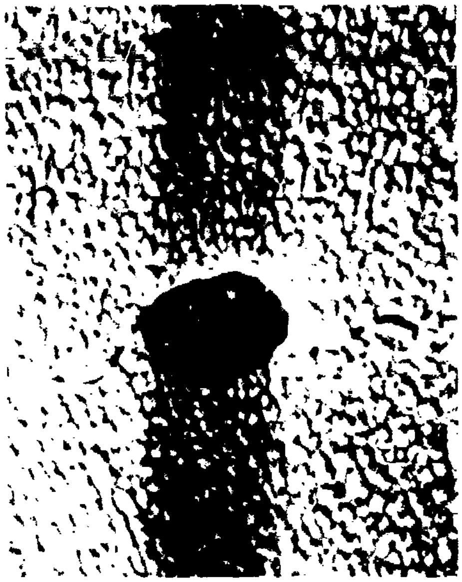
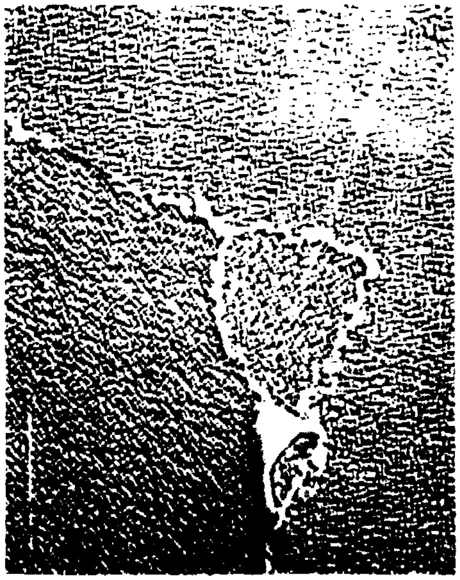


Fig. 10

390 X
A. Engine Run & Conventional Process



2700 X
B. Engine Run & Conventional Process



2100 X
C. Engine Run + Hip Process

Fig. 12 Effect of Hip on Engine Run Rene 80 Turbine Blades

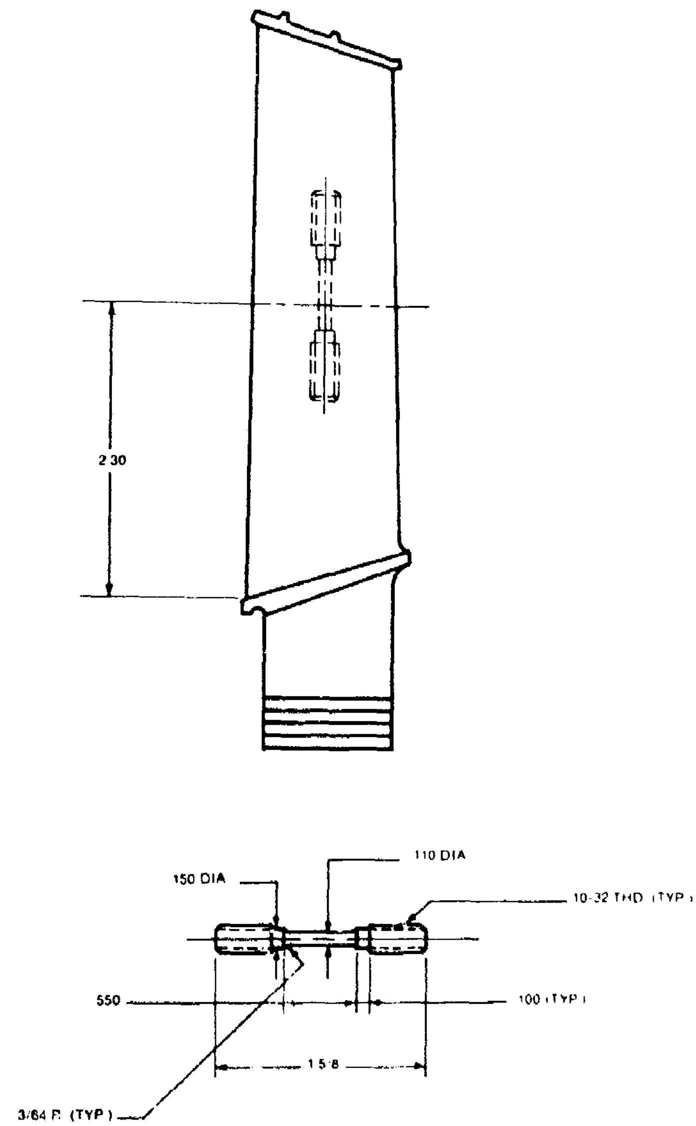


Fig. 13 Configuration of Test Specimen LP-1 Blades

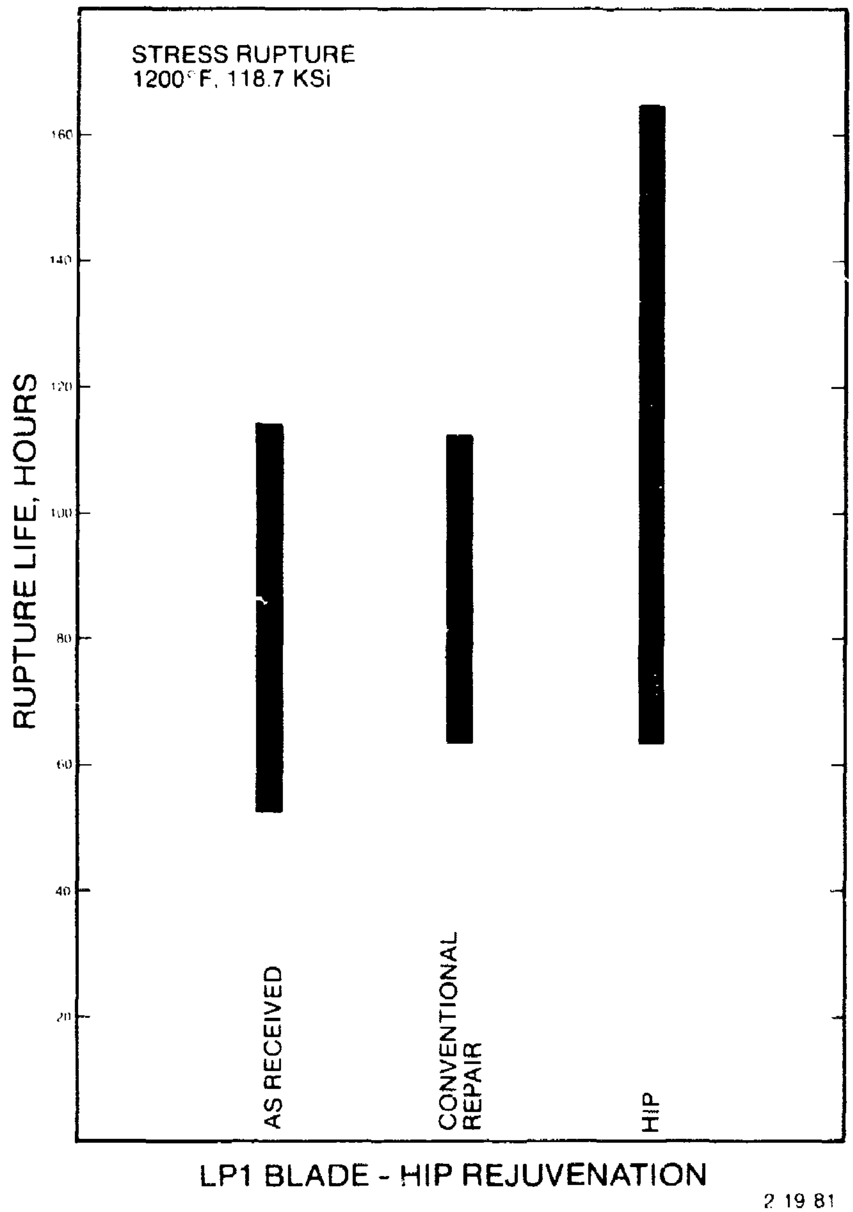
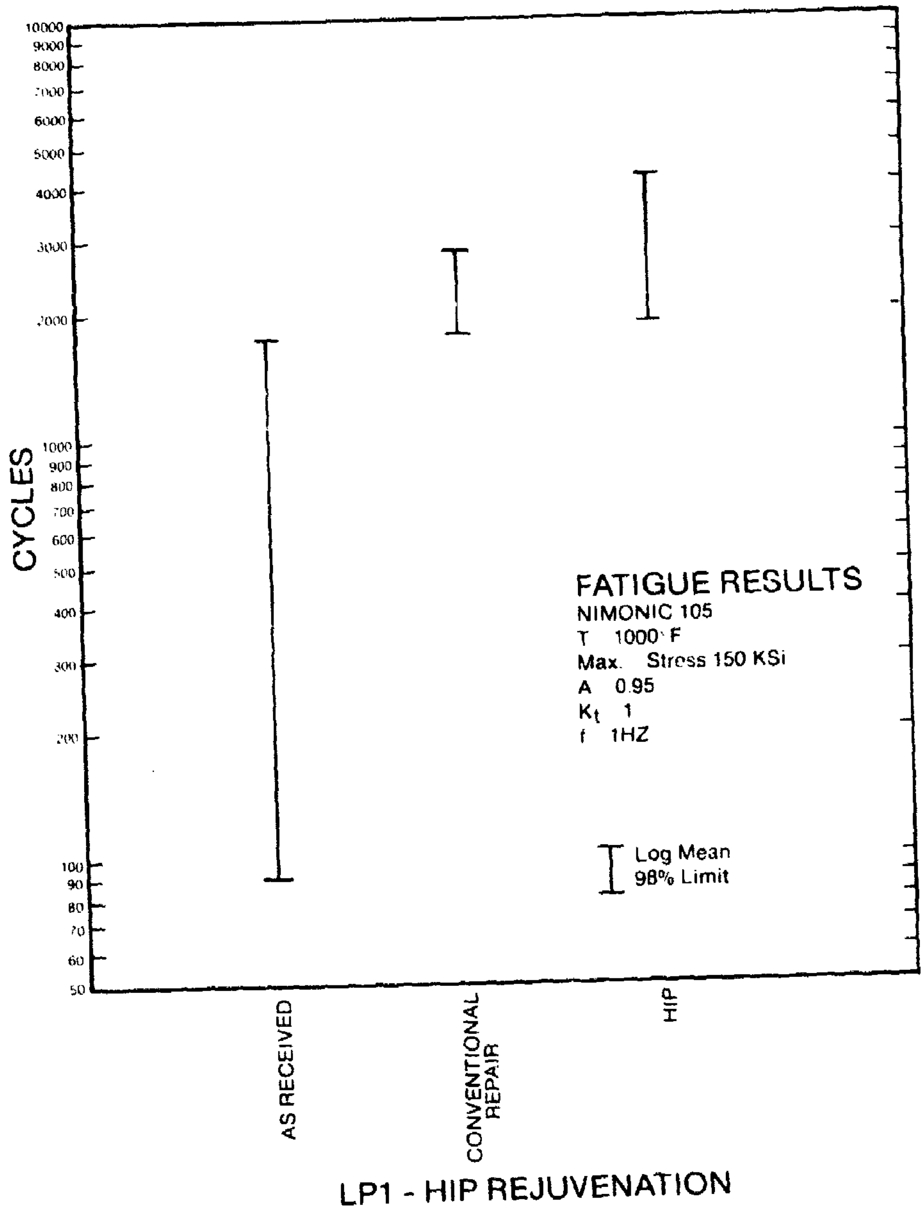

STRESS RUPTURE
1200°F, 118.7 KSi
RUPTURE LIFE, HOURS
160
140
120
100
80
60
40
20
AS RECEIVED
CONVENTIONAL REPAIR
HIP
LP1 BLADE - HIP REJUVENATION
2 19 81

Fig. 14



CYCLES
10000
9000
8000
7000
6000
5000
4000
3000
2000
1000
900
800
700
600
500
400
300
200
100
90
80
70
60
50
FATIGUE RESULTS
NIMONIC 105
T 1000° F
Max. Stress 150 KSi
A 0.95
K_t 1
f 1HZ
Log Mean
98% Limit
AS RECEIVED
CONVENTIONAL REPAIR
HIP
LP1 - HIP REJUVENATION

Fig. 15

Effect of HIP on Mechanical Properties of DS MARM-200 Alloy Components

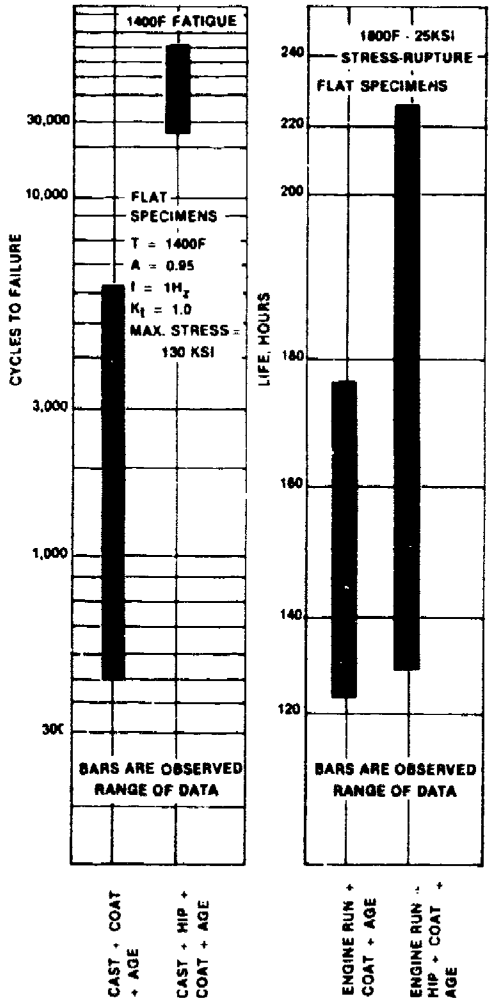


Fig. 16

RECORDER'S REPORT SESSION II

by

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The papers in the session "Life extension and repair II" could be brought under three different headings.

(1) Some papers, the papers which were presented by Mr Marijnissen of Elbar and Mr Pichoir of ONERA, dealt with high temperature coatings, their protection and the effect of coatings on the mechanical properties of the substrate.

(2) The paper given by Mr Honnorat of Snecma can be brought under the heading: Repair braze processes.

(3) The other two papers, given by Mr Leach of Westinghouse and Mr v.d.Vet of Chromalloy, discussed the rejuver :tion of turbine components by the use of the HIP process, which means Hot Isostatic Pressing.

Although all the above processes do not seem directly related to each other, they have several things in common. Firstly, they are mostly applied on high temperature turbine parts. Secondly, and most importantly, they all serve the same purpose: the extension of service life of high temperature parts which are very, not to say extremely, expensive. To give an example: present 1st stage turbine blades will cost up to about 2000 dollars each; vane segment prices can be even higher.

Let us go back now to the first subject in this session: the coatings. Mr Marijnissen of Elbar told us about the Elcoat 360 coating, which is a Ti-Si coating developed by his company for use on industrial gas turbines. The coating is obtained by firstly applying a thin titanium layer on the surface by means of ion-plating, followed by a diffusion heat treatment. Subsequently the part is pack-siliconized and finally aged The coating application heat treatment is part of the heat treatment of the substrate. According to the speaker this Ti-Si coating should demonstrate a superior corrosion resistance at around 700°C and equal corrosion resistance at about 900°C in comparison to the MCrAlY type overlays. Unfortunately these comparative results were not included in the paper. Nevertheless, according to the published rig testing results this coating seems to be fairly promising, at least for the lower temperature range.

Mr Pichoir of ONERA gave a presentation about the effect of coatings on the mechanical properties of superalloy components. The adverse effects of coatings can be due to (1) the thermal cycle of the coating application treatment; (2) the reduction in cross section and the coating weight, resulting in higher loading of the net, load-bearing cross section; (3) the interdiffusion between coating and substrate during service, which might result in unfavourable precipitation and (4) the microstructure of the coating itself, e.g. a low ductility of the coating resulting in early crack initiation.

The author emphasised that dependent upon the nature of the coating (pack-aluminides in comparison to the overlays) the effect on the mechanical properties can be quite different. E.g., the overlay coatings show less negative effects on the mechanical properties because their thermal application cycle can easily be made compatible with the superalloy substrate heat treatment. Furthermore, the interdiffusion between overlays and substrate is only of minor importance owing to the better stability of the overlay coating. Finally the overlay shows much better ductility.

In the discussion on these papers the questions were especially directed to the ductility of the Elcoat 360 coating and the behaviour under thermal cycling. According to the author the ductility of the coating and its cyclic behaviour should be good.

The second subject in this session was repair brazing.

Mr Honnorat gave an interesting presentation about the technique of diffusion brazing for repair of cracked and even further degraded turbine stator components. He described two processes: (1) brasage diffusion, which is a diffusion brazing process using a low melting point braze filler metal, based on a Ni-Co or Ni-Cr base alloy with Si and B-additions. The braze filler metal can be applied in the form of a metal foil or as a paste containing braze metal powder. This process can bridge gaps up to a maximum of about 100 μm.

The second process the author mentioned was "rechargement brasage diffusion" which is also a diffusion brazing process. However, in this case the braze filler metal is always applied as a paste. Furthermore, apart from the low melting point braze powder a second type powder is added, which has the same or approximately the same composition as the base material of the component, and which will not melt during the process. This allows for the repair of wide gaps and even allows for the build up of worn-off and corroded surfaces. The described techniques, which seem to be very similar to the ADH process developed by GE, look very promising.

The final subject in this session was the potential of HIP processing on rejuvenation of turbine components. Both authors, Mr Leach of Westinghouse and Mr v.d.Vet of Chromalloy pointed to the beneficial effect of the combination of applied pressure and heat treatment during the HIP process.

During hipping the component is simultaneously subjected to high temperature and very high pressure. Such a treatment will on the one hand result in eliminating structural discontinuities (e.g. by closing up internal creep voids). On the other hand the process has the potential of reheat treating the component and reversing the microstructural degradation which has occurred in service. On several alloys hipping has shown really remarkable results, on other alloys hipping seems to have a negative effect. The most important problem still is the selection of a suitable heat treatment cycle for each individual alloy, because temperature has a marked effect on the dissolution of the γ' phase and other precipitates and on grain coarsening. Also the cooling rate at the end of the HIP process itself or during reheat-treatment cycles is a major factor. A more fundamental knowledge of the combined effect of heat and pressure seems to be necessary to gain the full potential from this very promising technology.

To conclude: New coatings are being developed; many repair processes are appearing, e.g. diffusion brazing with the option of adding extra parent metal powder to bridge broad cracks, or even to build up airfoil surfaces; and hipping is underway as a means to rejuvenate rejected turbine parts still waiting in stock for better times. All these processes are highly important, yet I missed something which hardly was spoken of during the presentations and discussions. I will mention two simple things:

(1) Is there anybody here who can tell me how you can determine the remaining life of a coating when a blade comes in for overhaul? I will make this clear: if a blade comes in and it still looks good, who is going to say that you can return it to service instead of recoating it? Because if the coating is going to fail shortly after returning it to service then the blade will not even be repairable at the next overhaul, owing to severe parent metal attack. Thus the remaining coating life has to be measured in some way.

(2) All repair processes like diffusion brazing and also hipping can only eliminate surface connected cracks with very clean surfaces. Hence cleaning is a major topic. Perhaps we can cover these topics during the discussion

REGENERATION OF THE CREEP PROPERTIES OF A CAST Ni-Cr-BASE ALLOY

by

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SUMMARY

The effectiveness of hot isostatic pressing (hipping) as a means of removing the damage which accumulates during creep has been investigated by testing specimens for times up to [illegible] of their rupture life (approximately [illegible]h or [illegible]h), applying a hipping treatment and then continuing the creep test, under the same conditions, to failure. A comparison of the properties of material tested to rupture without interruption and material tested to rupture after hipping at some stage during secondary or tertiary creep has shown that improvements of up to 55% in the rupture life can be obtained by the use of this regenerative treatment and that the creep resistance is also restored. The practical implications of the work relate to the application of regenerative treatments to extend the life of gas turbine components removed from service during periodic overhaul.

INTRODUCTION

A number of studies (1-4) have shown that hot isostatic pressing, ie. the simultaneous application of high temperature and high pressure, can significantly reduce microporosity in cast superalloys. It is well established for wrought superalloys that if severe surface damage by mechanical or corrosive effects is not predominant, failure occurs by the growth and coalescence of pores or cavities at grain boundaries within the material (5) as a result of creep. The possibility arises therefore of applying the hipping process to remove this internal creep damage thereby restoring creep performance. Experiments on testpieces and on rotor blades subject to creep deformation have confirmed that the original properties can be recovered to a significant extent (6,7) and the useful life thus extended. The present work was carried out to investigate the feasibility of using the same process to recover the properties of specimens of a cast superalloy, IN738LC, which was known to fail as a result of cavity formation (8). Previous work on this alloy had shown that after various times of exposure in creep conditions a hipping treatment either gave more reproducible values of rupture life for a given stress without any otherwise significant improvement in properties (5,6) or could result in increased rupture lives up to about 35% (9). These tests involved relatively short lives to rupture compared to components in service and a feature of the present work has been that low stresses were used to give lives to rupture of approximately [illegible]h. Since it was important to determine the limit of the amount of creep damage that could be allowed to accumulate before hipping became ineffective, creep tests have been interrupted at various stages prior to rupture.

EXPERIMENTAL PROCEDURES

Two batches of testpieces were used, identified respectively as DCH and SU. The specimens of the DCH series were machined from investment-cast blanks while those of the SU series were machined from cast "carrots". The DCH specimens had a more uniform and slightly larger grain size but extensive tests showed that the difference was not sufficient to affect the creep performance (10). The analysis of the alloys used in this work was within the limits for the normal commercial specification (4), the DCH material having an electron vacancy number of 2.36 and that of SU being 2.30; the standard commercial heat treatment of 2h/1120°C/AC + 24h/845°C/AC was applied.

High sensitivity creep tests were carried out at 850°C using testpieces of the following dimensions: parallel portion, 80mm; gauge length, 50.8mm; gauge diameter, 7.6mm. The gauge length was determined by circumferential ridges to which extensometer limbs were connected. All tests at 170MPa were carried out on DCH material and those at 250MPa on SU material. To study the effect of hipping on material with varying amounts of creep damage tests on different specimens were terminated after progressive amounts of creep viz. from early secondary to the tertiary stage of deformation. The testpieces were then given a regenerative treatment which involved hipping for 2h at 1180°C in a pressure of argon of [illegible]MPa followed by the commercial heat treatment. The surfaces were cleaned by removing about [illegible]mm from the diameter and each specimen was then retested to failure under the same conditions of stress and temperature as previously. Base-line data were obtained by testing two specimens of DCH material at 170MPa to fracture without interruption and similarly one specimen of SU material at 250MPa.

RESULTS

Tests on seven specimens of DCH at 170MPa were interrupted at the times and strains detailed in Table 1 and described as condition 'B'. After applying the regenerative treatment the tests on five specimens were continued to failure, but two specimens, viz. [illegible] and [illegible], were retained for metallographic examination with no further creep testing. Data for the specimen of SU material which received the same sequence of testing, regeneration and retesting, but using an applied creep stress of 250MPa, are also contained in Table 1. The life and strain to fracture for each of the specimens after hipping, ie. condition 'C', is detailed also in Table 1 and the combined life, ie. inclusive of that before and after the hipping treatment, is given by 'B + C'. The total life to rupture may be compared with that for tests to fracture without interruption in the same test conditions, ie. condition 'A' in Table 1, and it will be observed that for all tests except that stopped at the earliest stage ([illegible]) there was an increase in total rupture life and, to varying extents, in rupture ductility as a result of the regenerative treatment.

The significance of these results becomes more apparent when the duration of the initial creep prior to hipping, expressed as a percentage of the expected nominal life, is plotted as a function of the total

life obtained by combining the durations in creep prior and subsequent to hipping, also expressed as a percentage of the expected nominal life. A graph illustrating the relationship obtained is shown in Fig.1, and it is evident that a greater benefit in increased life was obtained for specimens tested to the later stages of creep prior to hipping. Examples of the creep curves obtained before and after the regenerative treatment are given in Figures 2 and 3, for tests interrupted during steady-state and tertiary creep respectively. In Fig.3 a creep curve from a specimen tested to rupture without interruption is also shown for comparison. Figs.2 and 3 and the data in Table 1 show that the creep resistance was restored by the hipping process and from Fig.4 it is evident that the greatest improvement in creep resistance was obtained for the specimens with the highest creep rate prior to hipping.

The results of experiments reported elsewhere (10) showed that cavities on grain boundaries in cast IN738LC were observed after 1500h in tests carried out at 850°C and [illegible]MPa, so that creep damaged grain boundaries would be expected to occur in all the tests performed in this study. A metallographic examination to investigate the effect of hipping on creep damage on all specimens was clearly impossible because of the destructive nature of the examination required, so that observations of the effect of hipping have been confined to two representative cases. Thus tests on two specimens, 7DCH5D and 7DCH7D, were stopped at approximately 4500h and on a further two specimens, 7DCH11D and 7DCH10D, at approximately 7500h (Table 1). Subsequent to the application of the hipping procedure to these two pairs of specimens, one of each was retested in creep while the other was sectioned for metallographic examination. There was no evidence of any intergranular cavitation or cracking in either specimen so that it is reasonable to assume that the creep damage had been healed as a result of the hipping treatment.

DISCUSSION

It is evident from the results presented that hipping can restore the creep performance of specimens crept to various stages of secondary and tertiary creep, presumably as a result of the restoration of the original microstructural condition (6,9). Except for the specimen hipped after less than 20% of the expected life, the total life to rupture was extended in every case and the increase in duration was more marked for the specimens which had been crept for the longest times prior to receiving the regenerative treatment. A feature of the results is that for the three tests interrupted after creep durations of between 3000 and 6000h (creep strains of up to 1.5%) the life to rupture after hipping was approximately the same ie. ~7000h. This may represent the maximum life recoverable by a single regenerative treatment. Significantly, for the single specimen crept to the beginning of rapid tertiary (2.6% strain), the life after hipping was approximately 1500h less than this apparently optimum recoverable life. Thus the average line drawn through the points in Fig.1, which assumes that there is no decrease in the efficiency of hipping at high prior-strains, may not be valid. However, further work would be necessary to establish the precise form of this relationship which could have important implications for the practical application of hipping techniques to the recovery of properties in components removed from service.

There was no metallographic evidence to suggest that hipping became less effective after higher creep strains since grain boundary cavitation and cracking was removed in the specimen crept to the tertiary stage. Also the creep resistance was restored suggesting that the morphology of the microstructure was similar to that in the "as heat-treated" condition.

Surprisingly the recovery of creep life in the specimen interrupted in the early stage of creep (0.5% strain) also showed less than the apparent maximum amount of recoverable life. This is illustrated by the dotted line in Fig.1 since it would be expected that any benefit of hipping after creep would result in a total life to rupture of more than 100%. One possible explanation is that the hipping operation was somewhat damaging. In this case also further work would be necessary to establish a complete understanding of the behaviour since in a limited programme the effects of variability of performance, characteristic of cast alloys, cannot be assessed.

Despite these limitations the work has shown, for the first time to our knowledge for a cast alloy under the conditions of test described (viz. low stress and long durations), that creep performance can be restored by a hipping treatment. In terms of the practical application the evidence that significant extension of life can be obtained by regenerative treatment applied in the later stages of creep deformation is obviously advantageous particularly from the point of view of planning maintenance intervals. However it will be important to identify the stage at which effective regeneration will not be achieved. For the present alloy, comparison with other work (10) suggests that full regeneration of properties is obtained up to the stage at which intergranular cracks are no greater than about 0.03mm in length. Thus, for this particular alloy, the observation of internal crack length may provide a simple criterion for assessing the probable efficiency of a regenerative treatment although parameters appropriate to the lower applied stresses encountered in practice may have to be established. Also no effort has been made to investigate the effects of successive hipping operations but the results from work on a wrought alloy suggest that further restoration of properties could be expected (11).

In the wider context of repair of components the use of a welding process may result in a locally heterogeneous microstructure and occasionally in the formation of small defects. The application of regenerative treatments which include hipping after weld repair will ensure the re-establishment of a microstructure with good creep resistance and rupture properties.

CONCLUSIONS

The regenerative treatment referred to in these conclusions is a combined process which involves hot isostatic pressing followed by commercial heat treatment and the conclusions relate specifically to the cast superalloy IN738LC.

1. The regenerative treatment was effective in recovering the rupture properties when up to 85% of the nominal life had been consumed.

2. The later the stage in creep at which the regenerative treatment was applied, the greater was the increase in total life to rupture. A maximum improvement of 55% in rupture life was obtained.

3. The maximum efficiency of the regeneration treatment, in terms of subsequent added creep life, is approximately constant over a prescribed but wide range of creep duration.

4. The creep resistance was restored for material in which the deformation had been allowed to proceed well into the tertiary stage of creep.

5. The minimum creep rate after the regenerative treatment tended to decrease as the creep rate at which the test was stopped prior to the regenerative treatment increased.

6. The practical implications of the work relate to the use of regenerative treatments to extend the life of gas turbine components removed from service during periodic overhaul.

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ACKNOWLEDGEMENTS

This work formed part of the European COST 50 Project on Materials for Gas Turbines. The hipping was carried out at FFV Linkoping, Sweden under the supervision of Mr. Y Lindblom who also provided valuable detailed advice and information. Assistance in the creep testing by Miss L. V. Game and in the metallographic work by Mrs M. S. Peck is also gratefully acknowledged.

TABLE 1

Influence of hipping treatment on creep properties of IN738LC at 850°C

Applied Stress MPa	Condition	Mark	Creep rate, % per h x 10^4		Duration h	Ductility	
			Minimum	End of Stage B		Elongation %	RA%
170	B	7DCH5D	3.3	3.3	1505-S	0.5	0.6
"	C	7DCH5D	2.5	-	5758-F	7.4	9.0
"	B + C	7DCH5D			7263	7.9	9.6
170	B	7DCH6D	2.9	2.9	3005-S	0.8	1.0
"	C	7DCH6D	2.7	-	7052-F	6.0	11.0
"	B + C	7DCH6D			10057	6.8	12.0
170	B	7DCH7D	2.26	2.9	4697-S	1.0	0.6
"	C	7DCH7D	2.69	-	7305-F	8.0	11.0
"	B + C	7DCH7D			12002	9.0	11.6
170	B	7DCH8D	2.14	2.2	4518-S	0.95	0.4
170	B	7DCH9D	2.37	4.5	6043-S	1.5	0.5
"	C	7DCH9D	1.3	-	7538-F	6.6	11.0
"	B + C	7DCH9D			13581	8.1	11.5
170	B	7DCH10D	2.4	8.5	7555-S	2.6	2.6
"	C	7DCH10D	1.7	-	5733-F	4.0	8.9
"	B + C	7DCH10D			13288	6.6	11.5
170	B	7DCH11D	2.2	9.4	7540-S	3.0	2.2
170	A	Mean of Two Tests	2.35		8680-F	5.25	7.0
250	B	SU01	21.0	30.8	553-S	1.5	1.3
"	C	SU01	23.5	-	742-F	5.0	11.0
"	B + C	SU01			1295	6.5	12.3
250	A	SU10	24.6		860-F	4.5	10.0

F = time to fracture S = test stopped at time given

Condition A = Cht + creep to rupture (Cht = 2h/1120°C/AC + 24h/845°C/AC)

Condition B = Cht + partial life creep prior to hipping
(hipping = 2h/1180°C/170MPa + Cht)

Condition C = Creep to rupture after hipping

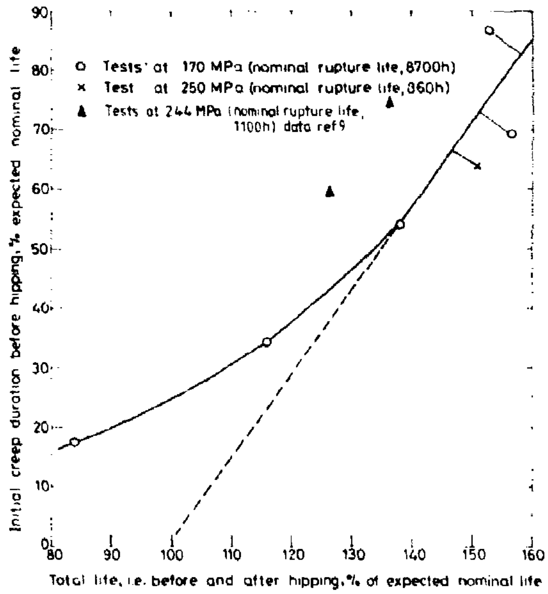

Fig. 1 Extension of rupture life of IN 738LC by use of a regenerative hipping treatment after different durations of creep prior to expected nominal rupture in the commercial heat-treatment condition.

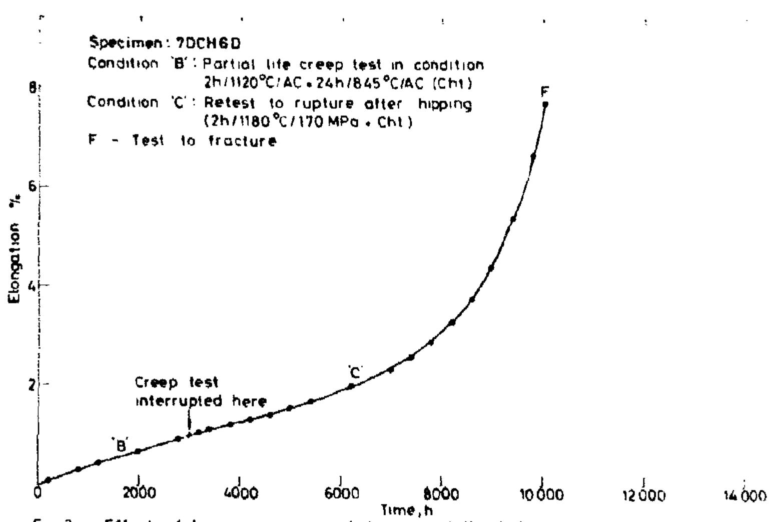

Fig 2 Effect of hipping on creep behaviour of IN738LC in test stopped during steady state creep at 170MPa and 850 °C

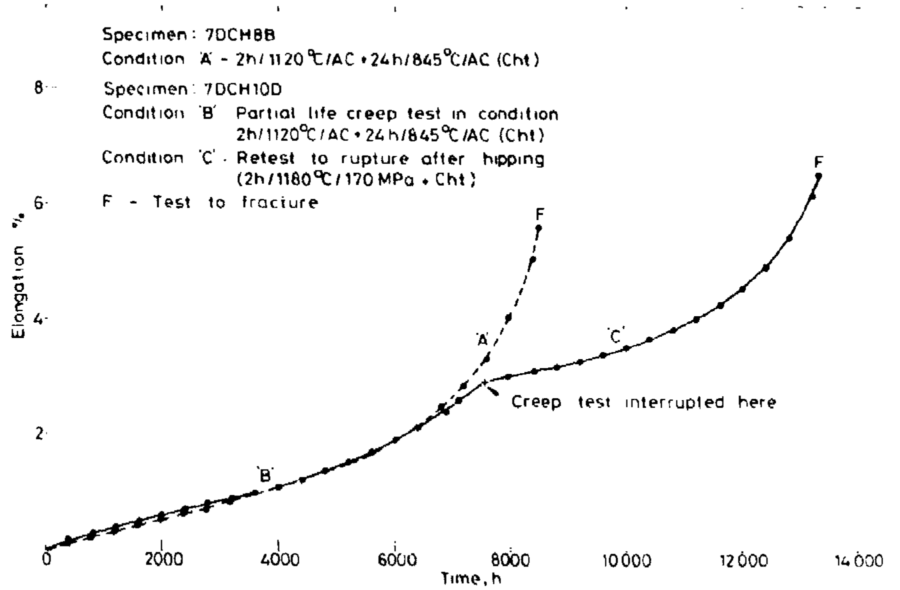

Fig.3 Effect of hipping on creep behaviour of IN738LC in test stopped in tertiary creep prior to rupture at 170MPa and 850 °C

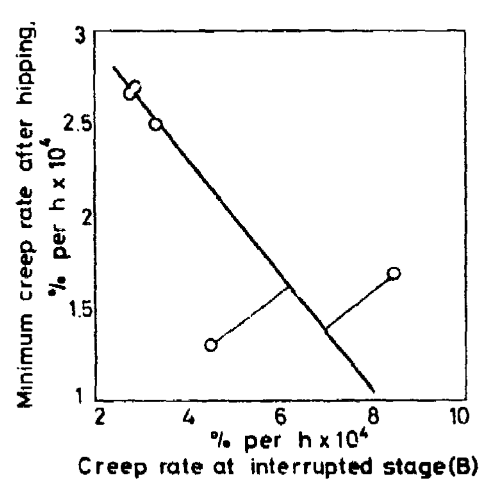

Fig. 4. Effect of creep rate reached at end of initial test on minimum creep rate after hipping treatment.

Repair and Regeneration of Turbine Blades, Vanes and Discs

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West Germany

Summary

Even today, the repair of incipiently cracked turbine components is essentially limited to non-rotating parts. Repairs to rotating parts are carried out on low-stressed areas, such as seals, only. In this case, weld build-up has proved to be a suitable process. Stator vanes can be high-temperature brazed following reduction annealing. However, problems are encountered when it comes to making sure of the complete removal of oxides. A highly promising method for increasing the reliability of turbine blades that have been in service lies in their regeneration by heat treatment or HIP-processing. Results to date have been so positive that one may reckon with the use of regenerated blades in the near future. However, a prerequisite for the use of these repair procedures is a guarantee of reliability and a knowledge of the stresses that occur during operation.

Repair of Turbine Blades, Vanes and Discs

Turbine components suffer from different kinds of loading, which in the highly stressed parts results in life-limiting consumption by creep, thermal fatigue and low-cycle fatigue. Several repair procedures have become well established for non-rotating parts in particular. Repair of rotating components, discs and blades, for instance, is usually restricted to lowly stressed areas, especially sealing fins, blade tips and couplings (Tables).

Welding and brazing techniques have been successfully developed even for materials that are difficult to weld, such as Ni-based superalloys with high γ' -content. Regenerative treatments have been specified and will be in use soon. This applies to turbine blades, and investigations have been started for disc materials.

In any case quality assurance methods have to be available to attest that the repaired component, base material and zone meet the specific strength requirements. Component tests under near-to-service conditions as well as test runs will in most cases be necessary to validate the repair procedures.

Repair of abraded seals

Because of unbalances and during running-in, seals of discs, shafts and blades are usually abraded, resulting in a decrease of engine efficiency. Several repair methods have been developed.

Figure 1 shows for example a detail of a repair instruction for turbine blades with seal tips at shroud and platform. These are removed completely by grinding the assembled rotor. A mating ring is prepared, sectioned, joint by brazing and finally the fins are restored (Fig. 2). The brazing temperature is about 1200 °C, well above the solution temperature and near the incipient melting point. Embrittlement of the grain boundaries by borides can also be observed, which may be detrimental to the thermal fatigue behaviour of the fins. Temperatures well above 1000 °C have to be used because of poor wetting behaviour of these alloys at lower temperatures (1).

To overcome the problems resulting from the heat treatment of the whole blade at highest temperatures a micro-plasma spraying or welding process would be preferred.

Figure 3 shows an experimental TIG welding, using blade material IN 100 and welding filler material Nimonic 90, which shows a micro-fissure-free heat-affected zone, despite the poor welding properties of IN 100. Thermal fatigue tests proved that this repair procedure produces sufficient properties. Stellite filler materials will further improve the wear properties of the sealing fins.

Abraded blade tips can be repaired using the same procedure (Fig. 4) (2).

Vanes

Crack formation in vanes is very common, in particular at trailing edges and near changes in section, owing to thermal fatigue-type loading. Because of the high temperature gas environment these cracks reveal heavily oxidized surfaces. Because of their thermodynamic stability, sophisticated methods have to be applied to reduce the oxides. Soaking under hydrogen or more effective fluoride atmosphere is used (3).

Problems arise particularly fom narrow cracks which have proved difficult to deoxidize and inspect. Incompletely cleaned crack surfaces will not be wetted by the brazing filler and consequently will give rise to crack propagation (Fig. 5).

It would seem necessary to monitor the repair process by metallographical examination of a representative sample.

Regeneration of Turbine Blades

Problem

The remaining creep life of a set of turbine blades has to be assessed on the basis of several microsections taken from samples. The criterion for the degree of consumption of creep life is the creep voiding of the grain boundaries. Classification of the sectioned blades is done with reference to a standard, which is assumed to give a correct correlation between consumption of creep life and the degree of voiding (Fig. 6).

The mean creep life damage of the test sample is taken as representative of the whole set and the remaining service time is assessed accordingly.

This procedure is unsatisfactory because of

- inspecting a relatively small test sample
- difficulties in correlating a definite microsection with a standard
- taking mean sample creep damage to define the remaining service time despite relatively large scatter.

We therefore feel that the probability for premature turbine blade failures associated with this process should be diminished by a suitable method, which could be a regeneration of the blades.

The blade in question, made of wrought Ni-base alloy Nimonic 108 (20 Co, 15 Cr, 5 Mo, 5 Al, 1.2 Ti), is cooled and coated (aluminium diffusion coating). An increase in the usable service lifetime of at least 30 % is thought to be sufficient to give an acceptable standard of in-service failure probability.

Treatments promising successful regeneration of the creep properties are:

- heat treatment typical for this alloy and
- hot isostatic pressing

on condition that no loss of other mechanical properties of the blade is caused by the treatment.

This means that the impact and fatigue properties must be examined closely after regeneration.

The investigation has been performed on test bars and blades. The bars were precrept up to about 75 % of their creep life, treated for regeneration and loaded finally to fracture. Blades were taken from engines, treated and tested in a hot gas test facility and compared with both new blades and those that had been in service.

In addition, fatigue (HCF), tensile impact tests and metallographic investigations were carried out.

Typical heat treatment

The first approach to restore the creep properties of samples and blades was the application of heat treatment typical for Nimonic 108:

$$4\ h/1150\ ^{o}C\ /\ AC + 16\ h/1030\ ^{o}C\ /\ AC + 16\ h/700\ ^{o}C$$

Results:

Creep properties of precrept bars (up to 70 % rupture time) were shown to be completely restorable (Fig. 7). More severely elongated bars (>7 %) were affected positively, revealing some ten per cent increase in total life.

Elimination of creep porosity is achieved by sintering and grain boundary migration during the solution heat treatment. Complete removal of voids was not attained (4).

Grain coarsening and precipitation of large inter-metallic phases within the coating occurred simultaneously.

HIP Treatment

Among others, the following HIP parameters were tested.

1 h / 1150 °C / 1600 bar

1 h / 1050 °C / 1800 bar

1.5 h / 1070 °C / 1320 bar

Partial heat treatments at 1030 °C and/or 700 °C were carried out additionally, the complete heat treatment showing the best results.

Results:

Total rupture life of bars and blades - already damaged in a test rig corresponding to about 75 % of their lifetime - was improved by at least 50 %.

No significant difference in the high cycle fatigue behaviour (bending fatigue) of virgin and HIP-treated blades and no embrittlement were observed.

Even subsurface grain boundary cracks can be eliminated (Fig. 8 and 9). No grain boundary coarsening or degradation of coating were observed at HIP temperatures lower than 1100 °C. The standard deviation of untreated blades is $s_{untr.}$ = 75 h, of treated $s_{treat.}$ = 40 h (Fig. 10).

The microstructure (γ' and carbide morphology) is restorable, with one peculiarity, that γ'-precipitates after HIP treatment are of cubic modification whereas the normal modification is spherical. (The same is true after an additional heat treatment.) The explanation for this behaviour is probably a change in the misfit of γ and γ' under HIP conditions to lower values as well as a change in solubility of the alloying elements.

Conclusion

The creep properties of damaged test bars and blades proved to be restorable - at least partially - by both typical heat treatment and HIP treatment. But grain coarsening and degradation of the aluminium diffusion coating precludes the application of solution heat treatment.

A HIP treatment at a temperature level of about 1050 °C and pressures of more than 1000 bar would appear to be a successful method for restoring the service properties of turbine blades for more than 50 % of their designed life time.

Quality assurance has to guarantee closely-kept HIP parameters such us temperature, pressure, cooling rate and contaminants.

The application of a regeneration treatment will only decrease the probability of an engine failure, not prevent it, of course. For there is as yet no reliable method for sorting out those blades which suffer from internal cracks starting from the cooling holes. Even high definition X-ray radiography has not proved suitable. Because the probability of internal cracks of critical sizes is significantly small, these defects would seem to be tolerable.

In 1982 regenerated blades will be fitted in three flight engines, and others will run in an endurance test.

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Type of defect	Type of loading	Repair procedure	Problems
– Fatigue cracking in bores, holes gears	LCF	Removal of areas possibly incipiently cracked by machining (regeneration)	Complete removal
– Fretting in bolt holes and grooves	HCF, LCF	Shot peening, removal	Parameter monitoring, crack detection
– Worn contact surfaces	Wear	Metal spraying	Adhesion
– Abraded seal lips	Wear, thermal fatigue	Weld build-up, micro-plasma spraying	Welding cracks, defects
– Worn grooves	Wear, HCF	Electroplated protection against wear	HCF strength, process monitoring

Table 1 **Repair Procedures for Shafts, Rings and Discs**

Type of defect	Type of loading	Repair procedure	Problems
– Coating damage	Thermal fatigue, erosion, corrosion	Mechanical, chemical stripping	Residues, wall thickness
– Creep damage	Creep-rupture	Regeneration	Effects on material, extreme damage to certain components
– Abraded seal lips	Wear, thermal fatigue	High-temperature brazing, weld build up	Welding defects
– Thermal fatigue cracks	Thermal fatigue	Reduction of oxides, high-temperature brazing	Cleaning of small cracks, effects on material
– Notches	FOD	Blending	Smearing of starting cracks

Table 2 **Repair Procedures for Turbine Blades and Vanes**

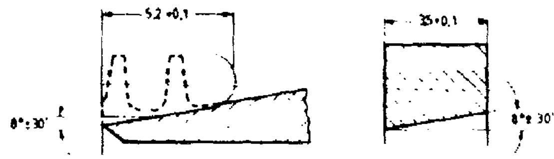

Figure 1 Detail of a Technical Repair Instruction for the replacement of abraded sealing fins by a high-temperature brazed piece of suitable material, final shaping by grinding

Figure 2 Microsection showing repaired sealing fins, blade material IN 100, brazing filler material René 80 + 2 % B

Figure 3 Weld build up of worn fins, TIG welding, filler material Nimonic 90, blade material IN 100

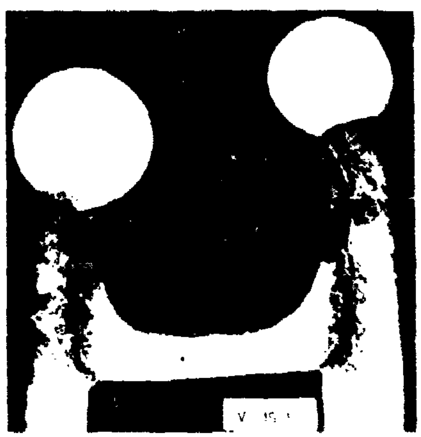

Figure 4 Weld build up of abraded blade tip

Figure 5 Repair of vane material by soaking in reducing atmosphere and high temperature brazing

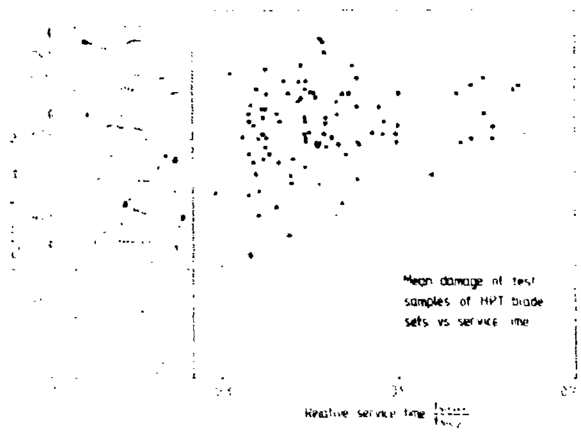

Figure 6 Consumption of creep life of sets of turbine blades vs. service time, as assessed by taking the mean creep damage of a test sample from each set

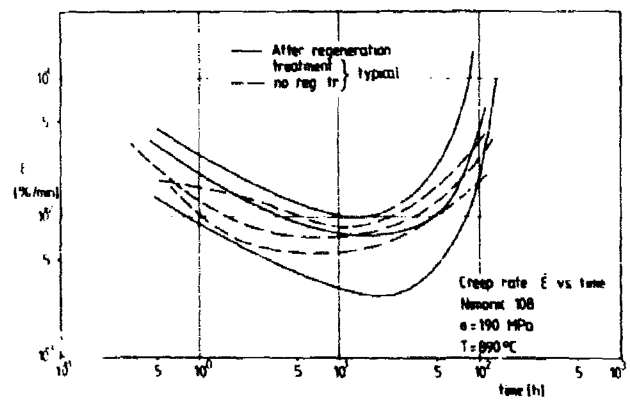

Figure 7 Creep rate of specimens with and without regeneration treatment

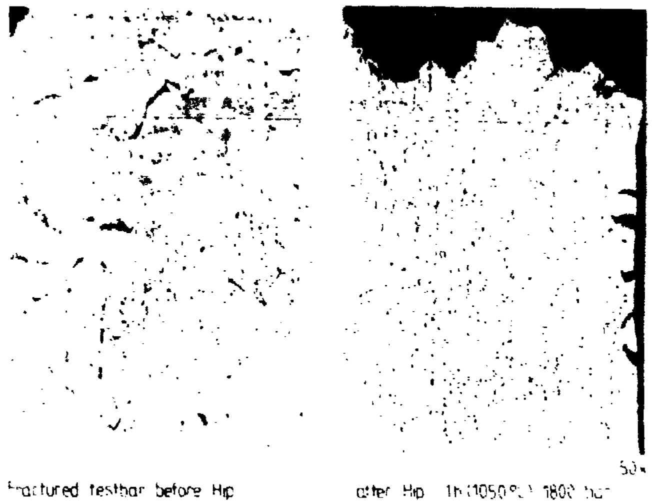

Figure 8 Grain boundary cracking in a creep rupture tested specimen before and

Before HIP (100x) After HIP (100x)

Figure 9 Microsections of blades before and after HIP treatment

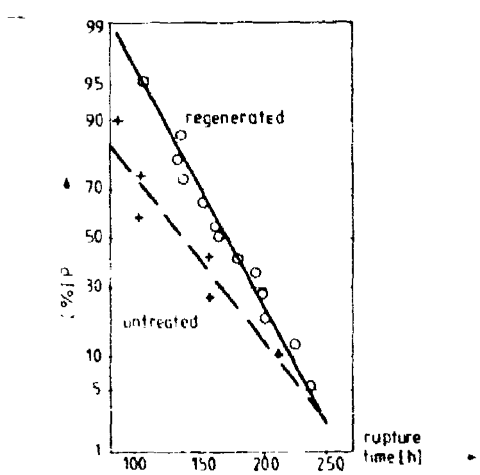

Figure 10 Rupture probability of blades tested in hot gas atmosphere, with and without regeneration treatment

A NEW APPROACH TO THE WELDABILITY OF NICKELBASE, AS-CAST AND POWDER METALLURGY SUPERALLOYS.

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SUMMARY

The investigation has shown that crack formation in the HAZ associated with the welding of high γ'-content nickelbase superalloys is due mainly to the shrinkage, which takes place during precipitation of the γ'. Crack formation can be prevented by controlling the cooling rate during welding of these alloys.
On the basis of measurements, which illustrate the relationships between a given cooling rate after welding, the partial effects of the elements Co, Cr, Al and Ti on the weldability were calculated, and this allowed a modified weldability diagram for nickelbase superalloys to be established. Simple hardness measurements appear to give really useful results for crack-free TIG welding, plasma welding, friction welding and EB welding of nickelbase superalloys. Small differences in chemical composition and the degree of homogeneity of the $\gamma - \gamma'$ structure can have a decisive effect on the welding behaviour of superalloys.

1. INTRODUCTION

The repair and recovery of both the stationary and the rotating parts of the hot section of a gas turbine is becoming more and more a generally-accepted practice. The regularly repeated assertion that repair of the hot section, and in particular the first stage inlet guide vane and rotating blades and buckets is not possible, is based more on commercial considerations than on any practical and/or theoretical basis.
In the special case discussed in this report, concerning the repair of nickelbase superalloys, such as are used in the first 2nd stages of the rotating sections of a gas turbine, welding plays an important part. Some types of damage, e.g. tip wear and FOD can be repaired economically only by welding. The application of welding for the repair of rotating parts, and in particular blades and buckets, is nonetheless affected by a number of factors, namely :

a) the stress level in the region of the blade, which is to be welded.
b) the temperature level in the region of the blade to be repaired.
c) the wall thickness of the region to be welded.
d) the material of the part to be welded.

As the wall thickness of the rotating parts to be repaired increases, accompanied normally by a higher level of stress, the weldability of the highly-stable nickelbase superalloys used for these parts falls rapidly.
The cobaltbase superalloys are generally characterized by much lower strength, but this is accompanied by much better weldability. Their use is generally confined to the stationary parts of hot section. A fortunate circumstance from the repair point of view is that an increase in the stress level from the tip to the root of a blade is accompanied by a corresponding decrease in operating temperature. The use of, for example, TIG repair welds has been hitherto mainly restricted by the following factors :

1) a lack of commercially available nickelbase superalloy filler wires of satisfactory high-temperature strength (creep characteristics) suitable for welding the highly stressed regions of blades.
2) the decreasing weldability as the wall thickness increases in the case of γ'-strengthened nickelbase alloys as used for the highly stressed regions of blades.

These two factors more or less balance each other. Up till now very little research has been carried out into the factors, which affect the weldability of high γ'-content nickelbase superalloys. It is moreover a fact that alloys, such as In 738, In 939 and U 720 were developed not with a view to their welding characteristics, but on the basis on high temperature properties, such as thermal stability, creep strength and hot corrosion resistance.

2. THE WELDING OF NICKELBASE SUPERALLOYS

2.1 Historical

One of the earliest publications in the field of the weldability of nickelbase superalloys dates from 1966 and provides a recognised model of HAZ cracking (Ref. 1). The greatest increase in practical repair work by Elbar, particularly in the case of stationary turbine hot section components, began in the last five years. Due to the necessity for repairs to an extremely difficult to weld superalloy, such as In 738 for example, the need arose for fundamental investigation of the factors, which affect the weldability of this alloy. The most important defects, which can occur when welding nickelbase superalloys, which contain γ' are HAZ cracking and PWHT cracking.
The investigation began 4 years ago under contract from the Ministry of Economic Affairs

within the scope of a European collaboration in the field of [illegible] welding, plasma welding, EB welding, laser welding, friction welding and diffusion welding of the alloys In 100, [illegible], and In [illegible] in the as-cast and P.M. condition.

[illegible] Results of the investigation

The investigation has shown that crack formation in the HAZ when welding high γ′-content high [illegible] [illegible] the great shrinkage which occurs as a result of the precipitation of [illegible] during cooling from the γ′ solvus temperature. [illegible] effects, such as local melting [illegible] some part. The investigation has also shown that by controlling the cooling rate after welding the greatest part of the γ′ precipitation can take place in the immediate neighbourhood of the weld pool.
The ductile zone between the weld pool and the precipitation [illegible] is well able to absorb the stresses arising but must have sufficient time to do so. In that case no stresses will arise, which are great enough to cause the material to fail (even in the event of any postweld heat treatment). The cooling rate for crack-free welding with the above named materials has been measured as a function of the welding speed and material thickness. Fig. 1 gives a hardness survey across the HAZ of a welded joint in In 738, made with TIG welding at various welding speeds and cooling rates. The peak hardness in the HAZ related to an optimal cooling rate [illegible] to a crack-free weld. Simulation of these cooling rates showed that for crack-free welding the γ′ in the HAZ must have a certain size. At his minimum size the hardness (and strength) of the material was exactly at its maximum. A long series of simulation tests has shown that this phenomenon occurs with both as-cast and P.M. superalloys.
Fig. 2 shows the relationship between maximum cooling rate and the size of the γ′-particles in the case of In 738. Fig. 3 is a typical cooling curve for In 738; max. cooling rate means the cooling rate from [illegible]. Fig. 4 shows the Vickers hardness related to a given γ′-particle size, as produced by the various cooling rates given in Fig. 3. The maximum hardness [illegible] optimal cooling rate at which a crack-free weld can be made.
Measurements were also carried out to discover the relationship between max. cooling rate, γ′ particle size and hardness in the case of In 738 as-cast, U 700 as-cast and P.M., In 100 as-cast and U 710 as-cast (see Fig. 5 and 6). Due to excessive inhomogeneity in the as-cast In 100 it appeared impossible to produce a useful hardness survey. It can be suggested that the size of the γ′-particles is a function of the maximum cooling rate, which is given by the following equation :

$$d^3 = \frac{C}{V}$$

where d = size of γ′ particles
C = a material-dependent constant
V = maximum cooling rate

From the measured maximum cooling rate against the sizes of the γ′-particles and the measured hardness against the sizes of the γ′-particles after welding, the following constants could be calculated for In 738, U 700 and U 710 :

C	In 738	=	68	(10^6 mm^3 °C/sec)
C	U 700	=	115	(10^6 mm^3 °C/sec)
C	In 100	=	3[illegible]	(10^6 mm^3 °C/sec)
C	U 710	=	51	(10^6 mm^3 °C/sec)

Putting C = (Al Ti Cr Co)
X Y Z V

Then :

	w/o Al	w/o Ti	w/o Cr	w/o Co
In 100	5,5	4,7	9,5	15,0
In 738	3,4	3,4	16,0	8,5
U 700	4,3	3,5	15,0	18,5
U 710	2,5	5,0	18,0	15,0

This gives 4 equations with 4 unknowns; the solution of these showed that the effects of the weight concentrations of the four elements on the growth rate of the γ′-particles were as follows :

X : Y : Z : V = 100 : 84 : -23 : -4,3
Al Ti Cr Co

It is clear that the effects of other elements could be calculated in a similar fashion.

In the case of In 738 the following "precipitation-rate numbers" were found :

(+340)	(+286)	(-448)	(-36,5)
Al	Ti	Cr	Co

Co is more or less indifferent, Cr is a strong inhibitor of γ′-precipitation, Ti is a <u>stronger</u> γ′ -former than Al ! (in atomic percentages).

<u>Note</u> :

Atomic percentages only are used for σ -safe calculations. Thus, a higher Ti-Al ratio can result in a higher rate of γ′ -precipitation without any effect on the Nv number.

When we graphically plot the compositions of a number of well-known superalloys, determined from empirically-derived formulas, a much clearer picture of the effect of the elements on the weldability is obtained. (See figure no. 7). This picture differs distinctly from graphs compiled by other investigators (ref. 3 and 4).

3. APPLICATION OF THE RESULTS OF THE INVESTIGATIONS TO WELDING PROCESSES OTHER THAN TIG- AND PLASMA WELDING.

The differences in the electron beam welding behaviour (crack susceptibility) between two apparently identical U 700 P.M. alloys can be satisfactorily explained on the basis of a difference in precipitation behaviour. Using the empirically determined effects of composition on the weldability diagram, the compositions of two U 700 P.M. types (U 700 P.M. A and B), which clearly differed in their welding behaviour were plotted on the diagram. This showed that the positions of these alloys were also distinctly different.
The more crack-sensitive U 700 P.M. type B was appreciably dispaced towards a region of higher crack sensitivity. In view of the fact that the measured sizes of the γ′-particles were not used in the determination of the empirical relationship in this case of the U 700 types mentioned, this results can be considered as remarkabl . Measurements made later and which illustrate the relationship between hardness, maximum cooling rate and γ′-particle size are presented in Figure 8 and 9. The more crack sensitive type B appears to produce a lower hardness for the same maximum cooling rate. On the basis of these results the welding parameters were modified in such a manner that maximum hardness was produced in the HAZ, resulting in a crack free weld. It seems therefore that there are two parameters which give an idea of the welding behaviour of high γ′ -content nickelbase superalloys. These are :

a) the compositional balance of Co, Cr, Al and Ti.
b) the hardness measured in relation to the cooling rate.

These two factors form in fact the basis for the development of a welding method with less crack-sensitive, γ′ -containing nickelbase superalloy filler wires with sufficient creep strength to allow welding of the highly-stressed regions of blades and buckets. The results of the investigation appear applicable to the friction welding of superalloys, but no further observations as regards this can be made here. Since preheating before welding obviously has an effect on the maximum cooling rate it would seem that crack-free welds can be produced by this means. For nickelbase superalloys, however, this is not really effective until the temperatures are above 600°C. Figure 10 shows the effects of several preheat temperatures on the hardness pattern in the HAZ of In 738. EB welding of nickelbase superalloys with preheat has been recently reported in the literature (ref. 2) without any rigorous theoretical explanation. Preheating for EB welding will, however, involve temperatures above 900°C.
An important conclusion, which also can be made is that pre-weld heat treatment can have an effect on the weldability of nickelbase superalloys only if it has a di[illegible] effect on the homogeneity of the γ′ - γ′ -structure. It is for that reason that p[illegible] weld hip treatment can sometimes have a positive effect on the weldability of the [illegible]icult to weld superalloys.

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   AGARD report - LS-91

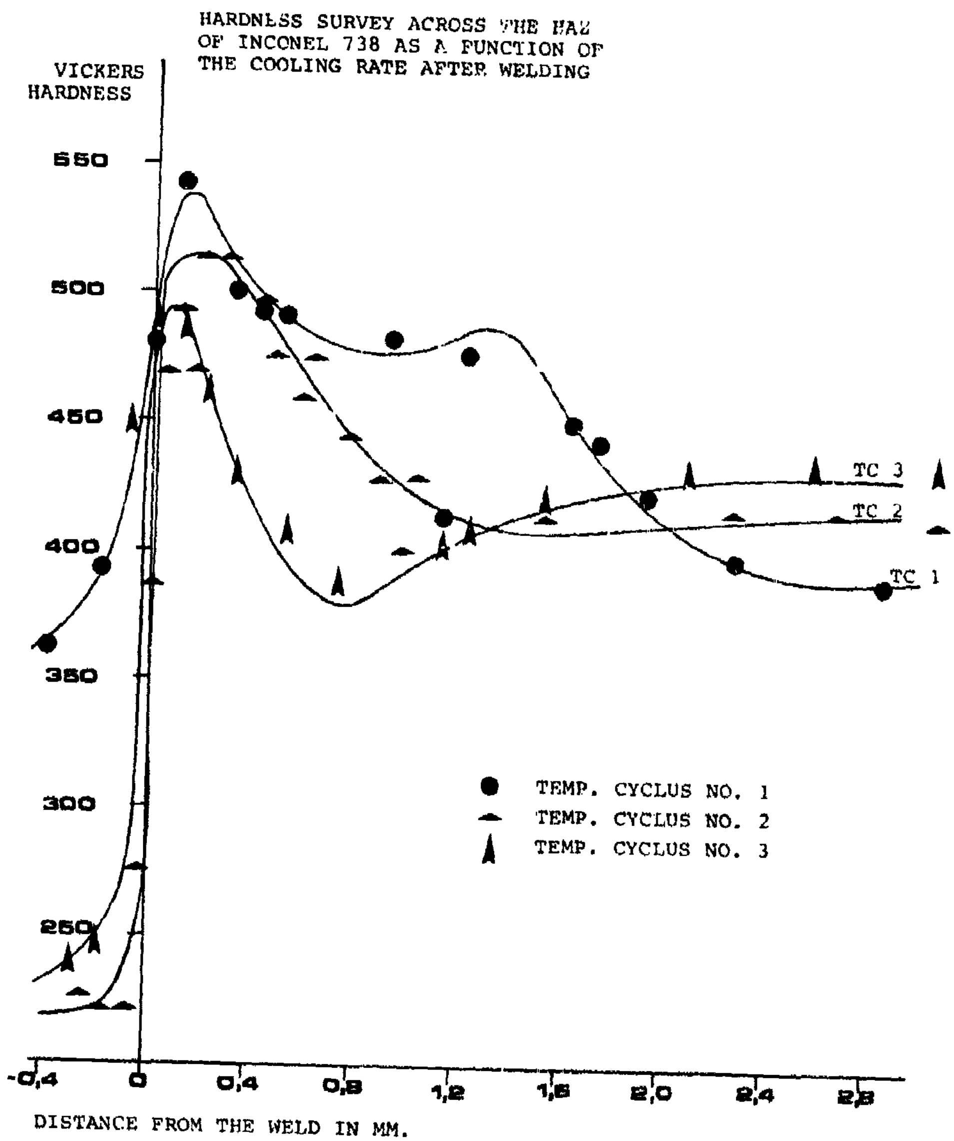

HARDNESS SURVEY ACROSS THE HAZ
OF INCONEL 738 AS A FUNCTION OF
THE COOLING RATE AFTER WELDING
VICKERS
HARDNESS
550
500
450
400
350
300
250
TC 3
TC 2
TC 1
TEMP. CYCLUS NO. 1
TEMP. CYCLUS NO. 2
TEMP. CYCLUS NO. 3
-0,4
0
0,4
0,8
1,2
1,6
2,0
2,4
2,8
DISTANCE FROM THE WELD IN MM.

Fig. no. 1

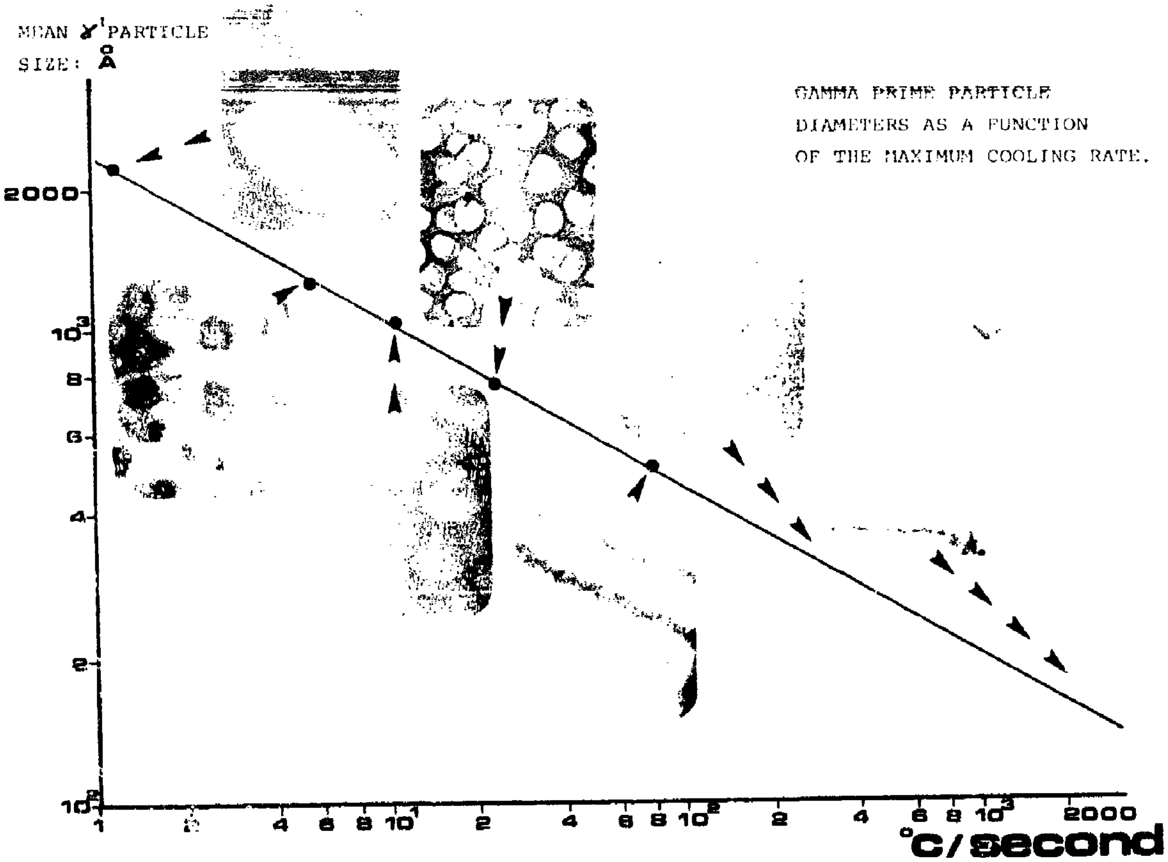

MEAN γ' PARTICLE SIZE: Å
GAMMA PRIME PARTICLE DIAMETERS AS A FUNCTION OF THE MAXIMUM COOLING RATE.
2000
10³
8
6
4
2
10²
1
2
4
6
8
10¹
2
4
6
8
10²
2
4
6
8
10³
2000
°C/second

Fig. no. 2

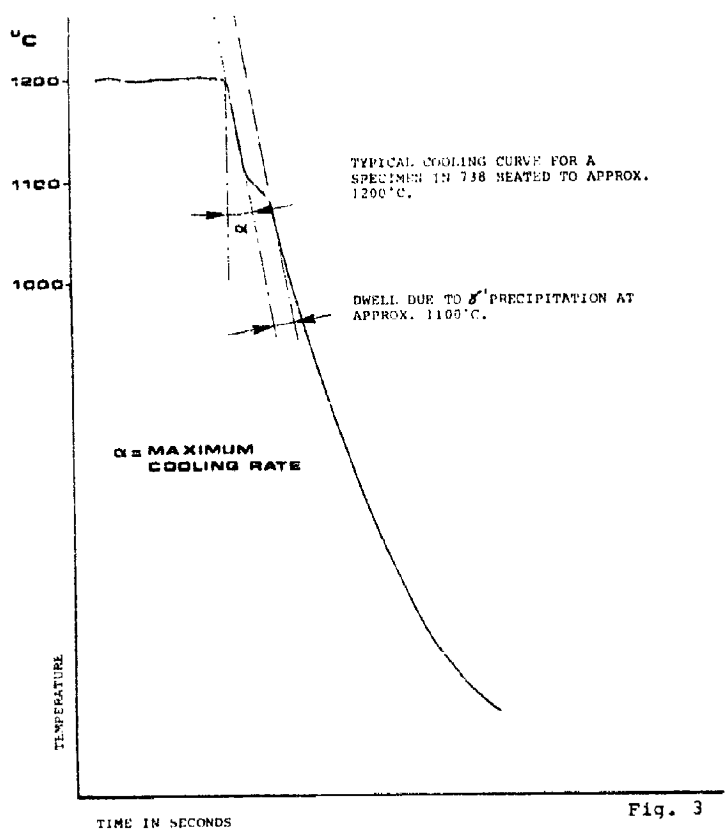
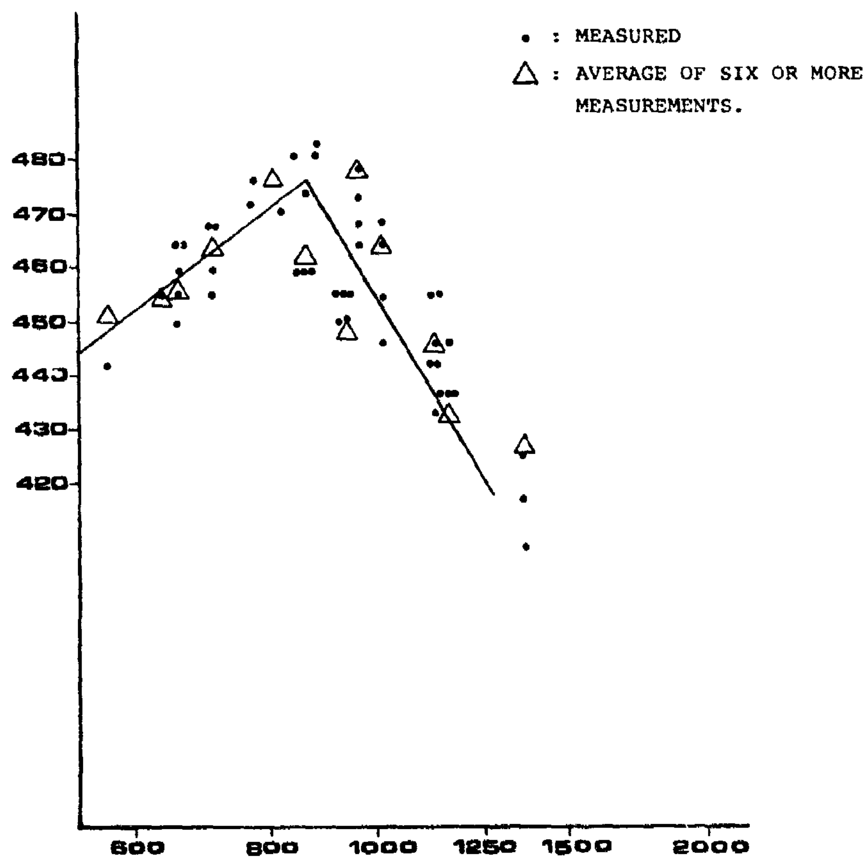

°C
1200
1100
1000
TYPICAL COOLING CURVE FOR A SPECIMEN IN 738 HEATED TO APPROX. 1200°C.
α
DWELL DUE TO γ' PRECIPITATION AT APPROX. 1100°C.
α = MAXIMUM COOLING RATE
TEMPERATURE
TIME IN SECONDS

Fig. 3

VICKERS HARDNESS OF INCONEL 738 AS A
FUNCTION OF GAMMA PRIME PARTICLE SIZE



GAMMA PRIME PARTICLE SIZE IN Å

Fig. no. 4

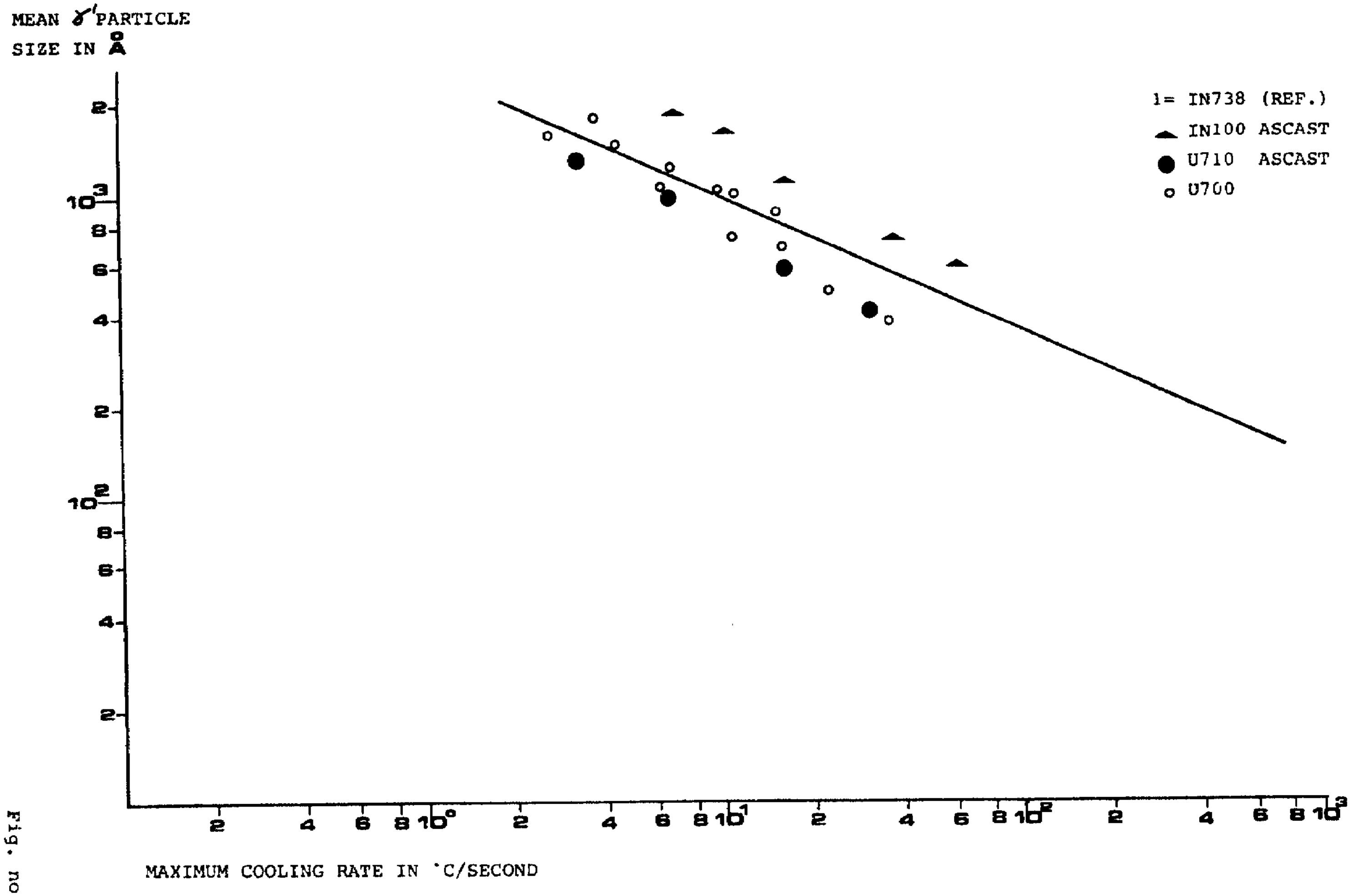

MEAN γ' PARTICLE
SIZE IN Å
1= IN738 (REF.)
IN100 ASCAST
U710 ASCAST
U700
MAXIMUM COOLING RATE IN °C/SECOND

Fig. no. 5

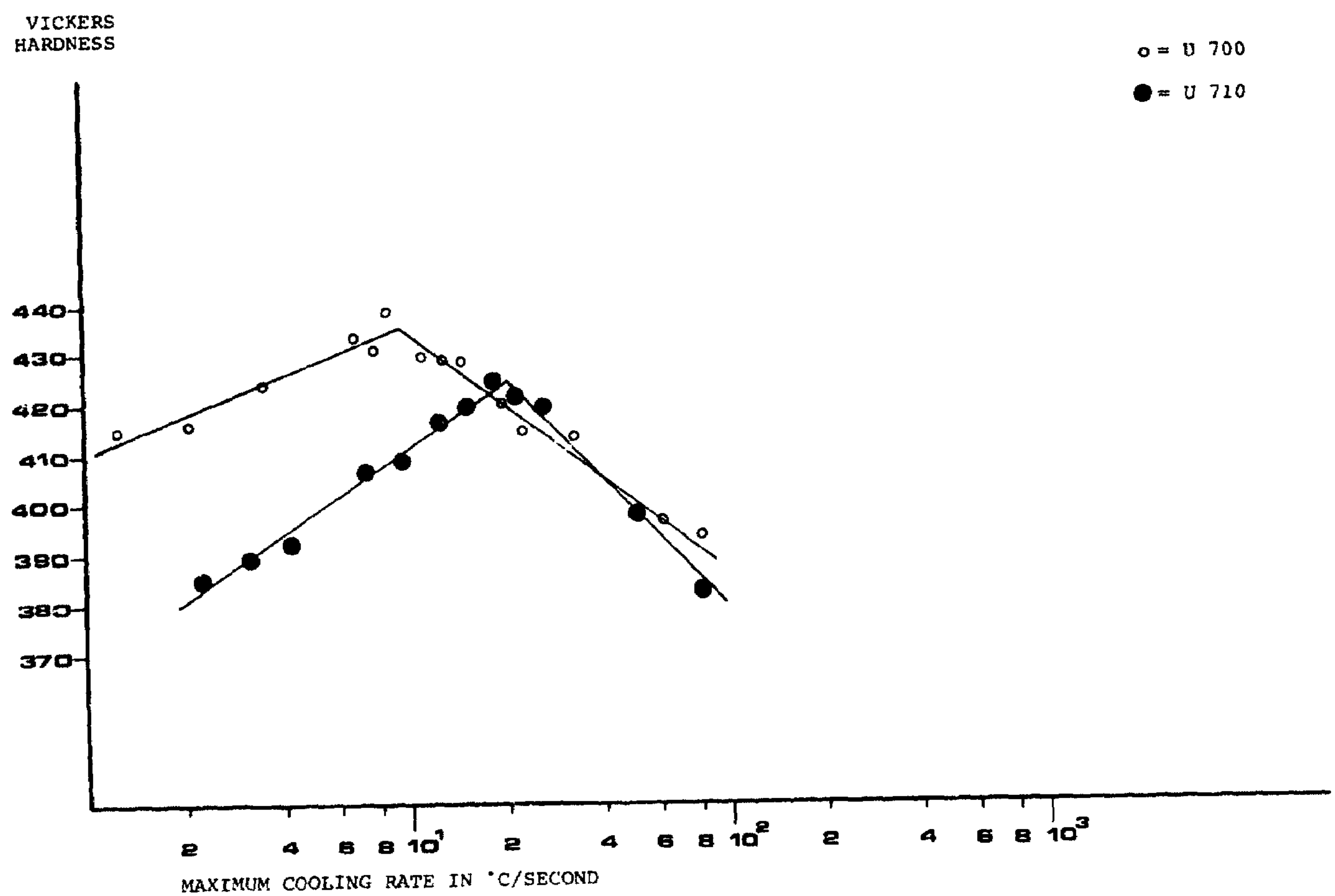
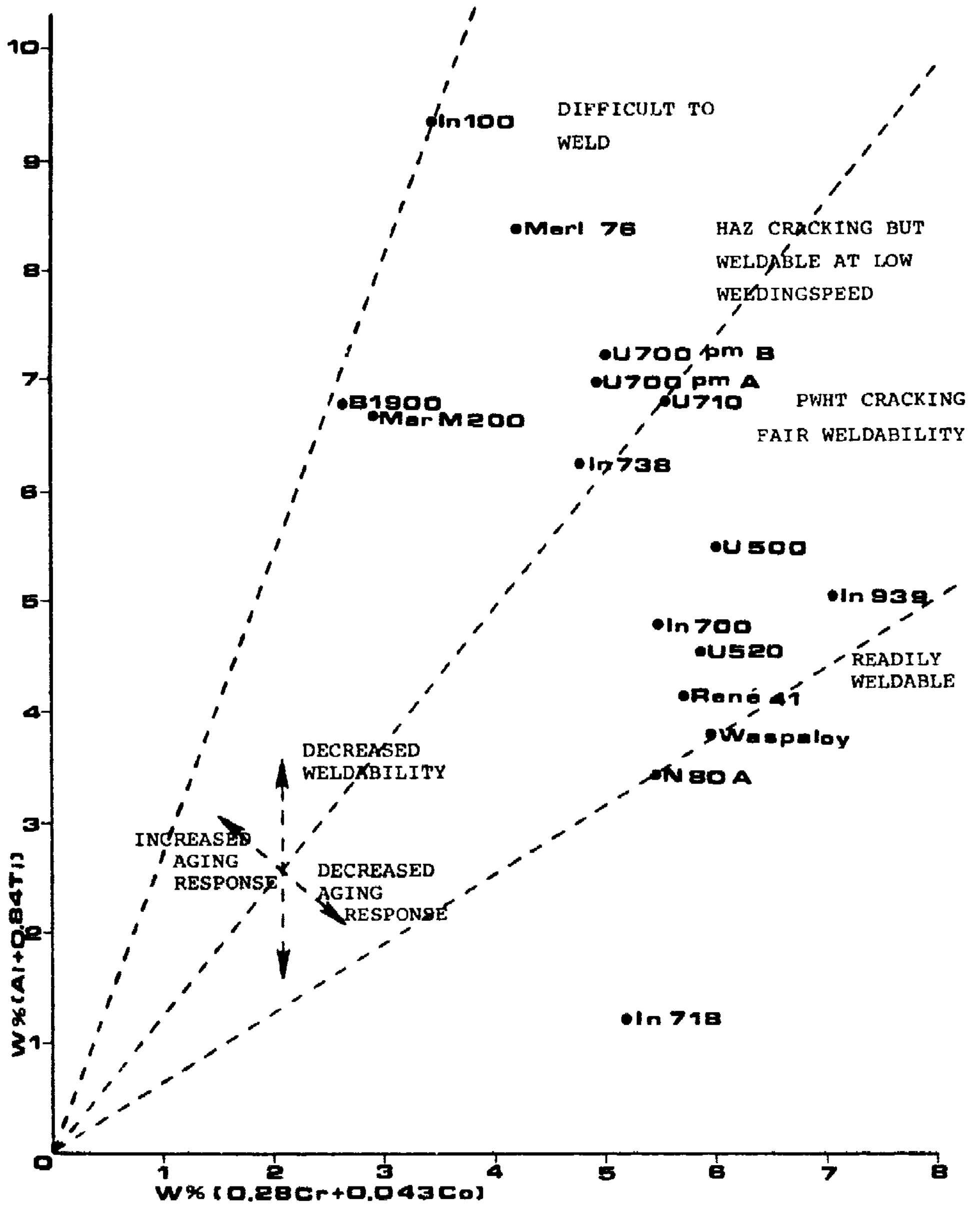

VICKERS
HARDNESS
o = U 700
● = U 710
440
430
420
410
400
380
380
370
2
4
6
8
10^1
2
4
6
8
10^2
2
4
6
8
10^3
MAXIMUM COOLING RATE IN °C/SECOND

Fig. no. 6



10
9
8
7
6
5
4
3
2
1
0
1
2
3
4
5
6
7
8
W% (Al+0.84Ti)
W% (0.28Cr+0.043Co)
In100
DIFFICULT TO WELD
Merl 76
HAZ CRACKING BUT WELDABLE AT LOW WELDINGSPEED
U700 pm B
U700 pm A
U710
PWHT CRACKING FAIR WELDABILITY
B1900
Mar M200
In738
U500
In 939
In700
U520
READILY WELDABLE
René 41
Waspaloy
N 80 A
DECREASED WELDABILITY
INCREASED AGING RESPONSE
DECREASED AGING RESPONSE
In 718

Fig. no. 7

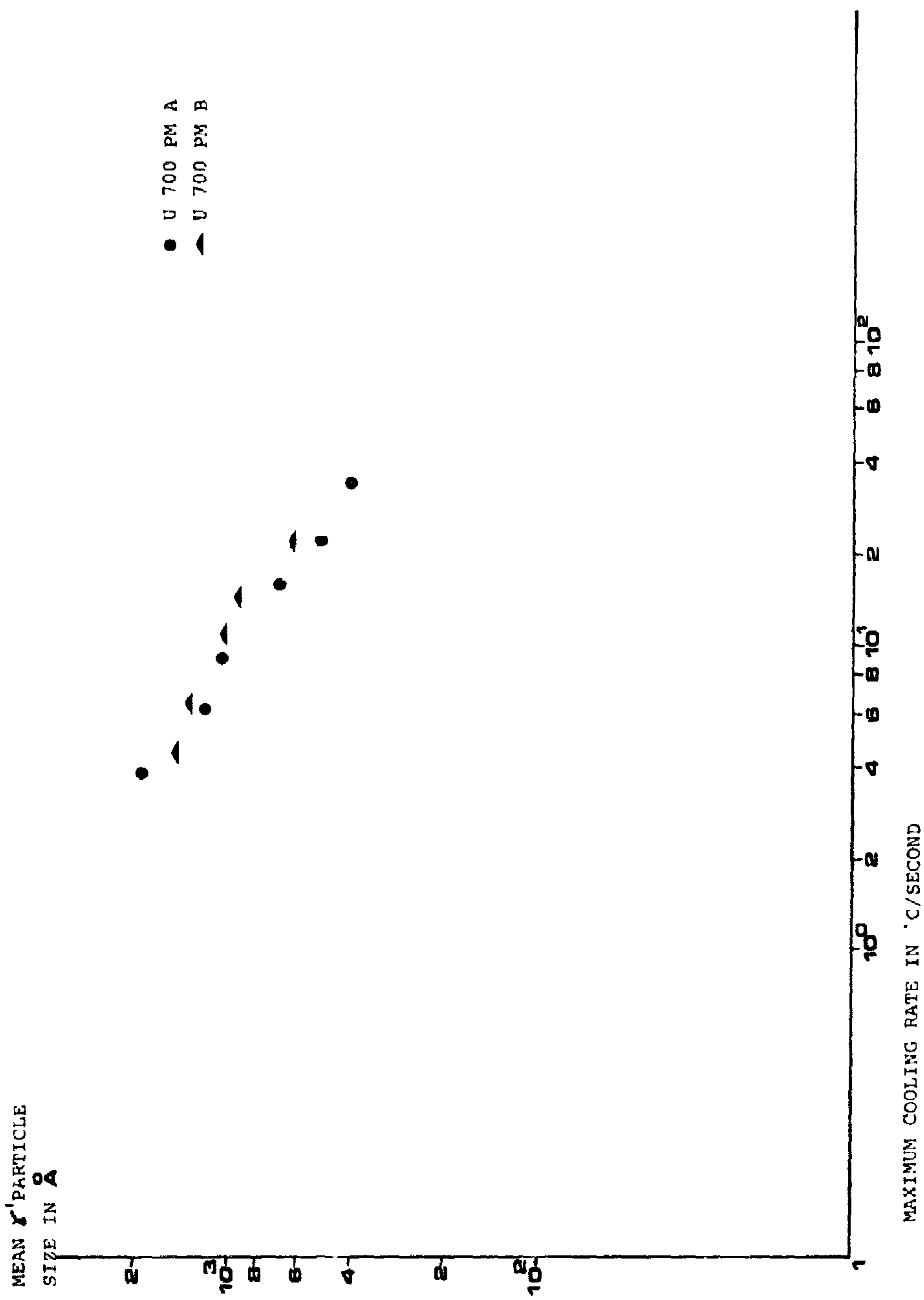
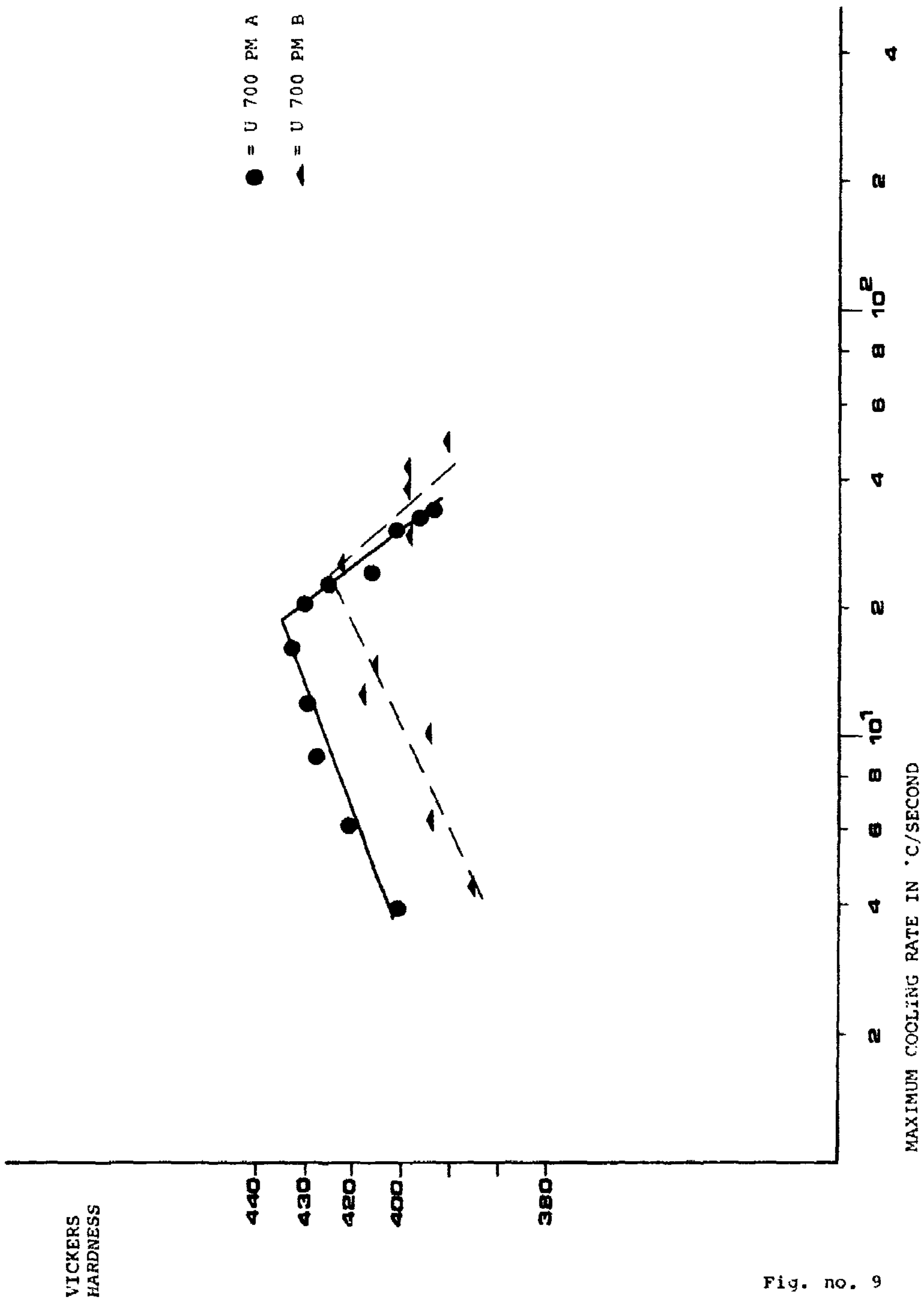

MEAN γ' PARTICLE
SIZE IN Å
MAXIMUM COOLING RATE IN °C/SECOND
U 700 PM A
U 700 PM B

Fig. no. 8



● = U 700 PM A
◄ = U 700 PM B
VICKERS HARDNESS
440
430
420
400
380
2
4
6
8
10^1
2
4
6
8
10^2
2
4
MAXIMUM COOLING RATE IN °C/SECOND

Fig. no. 9

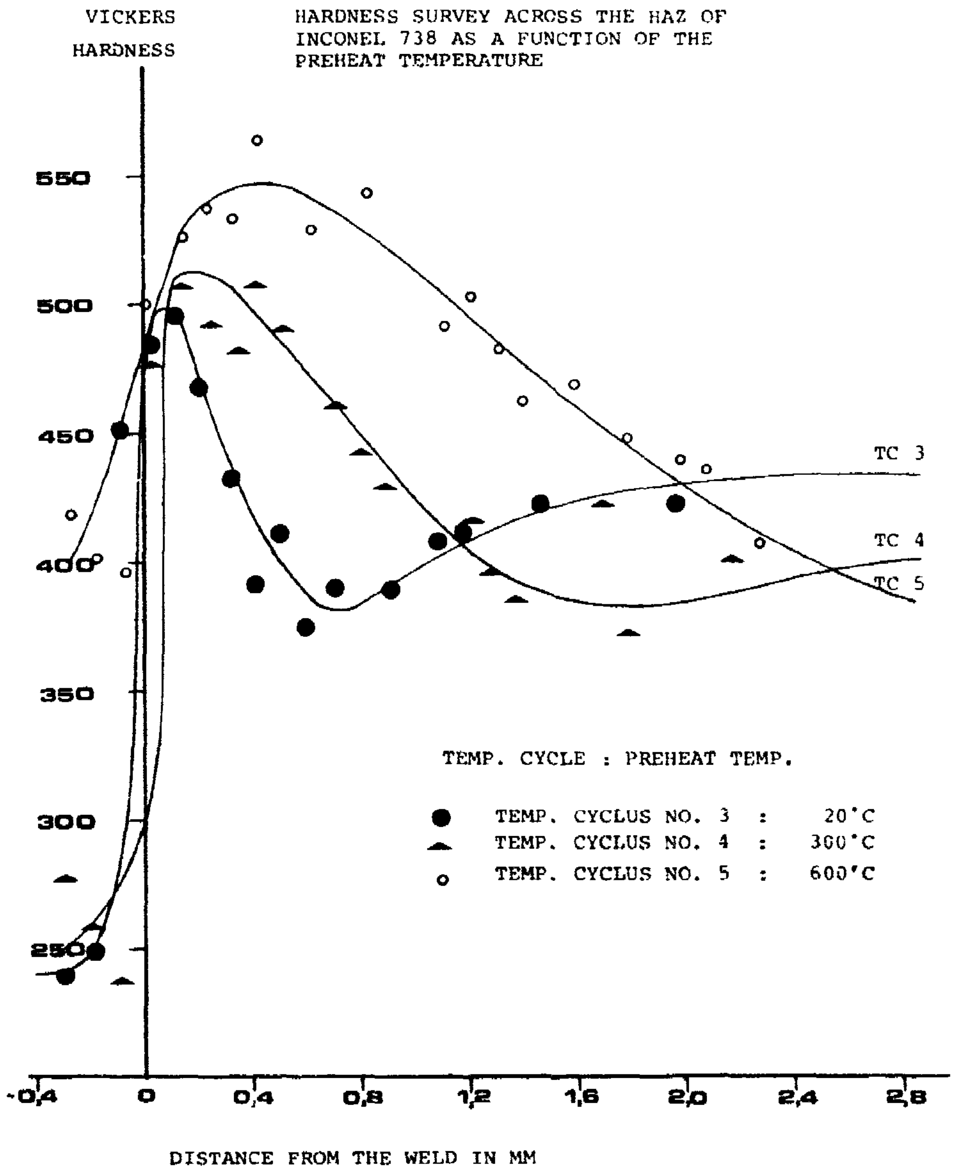

VICKERS
HARDNESS
HARDNESS SURVEY ACROSS THE HAZ OF
INCONEL 738 AS A FUNCTION OF THE
PREHEAT TEMPERATURE
550
500
450
400
350
300
250
TC 3
TC 4
TC 5
TEMP. CYCLE : PREHEAT TEMP.
TEMP. CYCLUS NO. 3 : 20°C
TEMP. CYCLUS NO. 4 : 300°C
TEMP. CYCLUS NO. 5 : 600°C
-0,4
0
0,4
0,8
1,2
1,6
2,0
2,4
2,8
DISTANCE FROM THE WELD IN MM

Fig. no. 10

RECORD'S REPORT – SESSION III

by

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The first two papers of this session continued from the previous discussion of hot isostatic pressing (HIP) to rejuvenate cast nickel base turbine blades. The third paper considered the weld repair of as-cast and powder metallurgy superalloys.

The first two papers showed the promising benefits due to HIP processing as a regenerative treatment to extend the life of gas turbines components removed from service during overhaul. The focus of these efforts were on the removal of creep damage during HIP processing. The importance of controlling the cooling rate after the HIP operation was emphasized. The following questions were discussed but not resolved.

(1) How many times can a turbine blade be rejuvenated?

(2) What is the effect of HIP rejuvenated on properties other than creep, that is, microstructure, fatigue behavior, etc.

The paper on weld repair of as-cast and powder metallurgy presented a framework for evaluating heat-affected-zone cracking as a function of composition and welding speed. The paper presented an interesting way to look at a very complex problem. It appeared that the data might be treated in a more fundamental way using thermodynamic, nucleation and growth and diffusion theory.

APPENDIX

COMMENTS ON THE MAINTENANCE IN SERVICE OF HIGH TEMPERATURE COMPONENTS IN AIRCRAFT JET ENGINES

by

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I first became acquainted with jet engines in 1946 and since that time I have, of course, seen many changes. Today's engines are far more powerful and fuel efficient; but they have lost one of their most beguiling features simplicity. Today's engines with their multiple stage turbines and concentric shafts are very complex and therefore much more difficult to maintain.

In 1946 military engines had a time between overhaul of 150 hours but very few of them attained even this time interval. The blade temperature was about 725°C and it has grown at the rate of about 15°C per year ever since. A truly remarkable achievement when all during this period metallurgists kept predicting that they had pushed alloys about as far as they could go.

It is perhaps more clear today that at any time in the past that the end is not in sight and I believe we should not only be working on extending and predicting the life of current engine components but we should devote considerable attention to making higher temperatures possible. It should, of course, be noted that in the past we wanted higher temperatures to enable aircraft to fly at higher speeds. Today we want these same higher temperatures so that the engine operates in a more fuel efficient manner. Some of the studies that I believe should be undertaken to prepare us for the future are listed below:

(a) We should seek a far more precise definition of the environment (temperature, stress, and chemical) than is available today. This definition would give us better clues to the mechanisms of failure and would enable the designers to come up with less demanding designs.

(b) We should develop far superior methods of keeping track of component histories so that we can better decide when rejuvenation procedures are appropriate.

(c) Surface crack cleaning still seems to me to be a major problem and better methods are needed before we can be confident of the repairs.

(d) I think there is further room for improvement in the presently used braze alloys. We are currently adding boron or other elements to reduce the brazing temperatures, however, it would appear to me that if we could find substitutes which would completely vaporize during the brazing processes, superior repairs could be made.

(e) New techniques are needed in both NDE (particularly for coatings and cooling passages) and welding. In the latter case I believe lasers should be looked at intensively.

(f) We should be getting ready for the newer materials that are already finding their way into jet engines. The principal one is, of course, conventional alloys made by powder metallurgical methods, single crystal blades, inter-metallic compounds and ceramics will gradually follow along. We should be prepared to inspect and maintain these new materials.

I would like to close with one final comment. We should also be very conscious of the gradually deteriorating quality of jet engine fuels. We already have seen a gradual increase in the percentage of aromatics these fuels contain and we know these can cause severe erosion problems and temperature increases in the engine. Other changes in fuel chemistry can introduce corrosive ingredients. We must ever be mindful of these changes in planning our maintenance procedures for jet engines.

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14. Abstract

All NATO countries need to combat the increasing cost of maintenance of engines and scarcity of strategic materials by improving component utilization. The objective of this Meeting was to review the problem areas and experiences in the maintenance of high temperature parts; many of these problem areas having a common base in relation to service experience and the characteristics of material behaviour, so that users may benefit from the advances in materials science and the future needs for R&D may be identified.

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